METALLURGIA

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(INCORPORATING THE METALLURGICAL ENGINEER)

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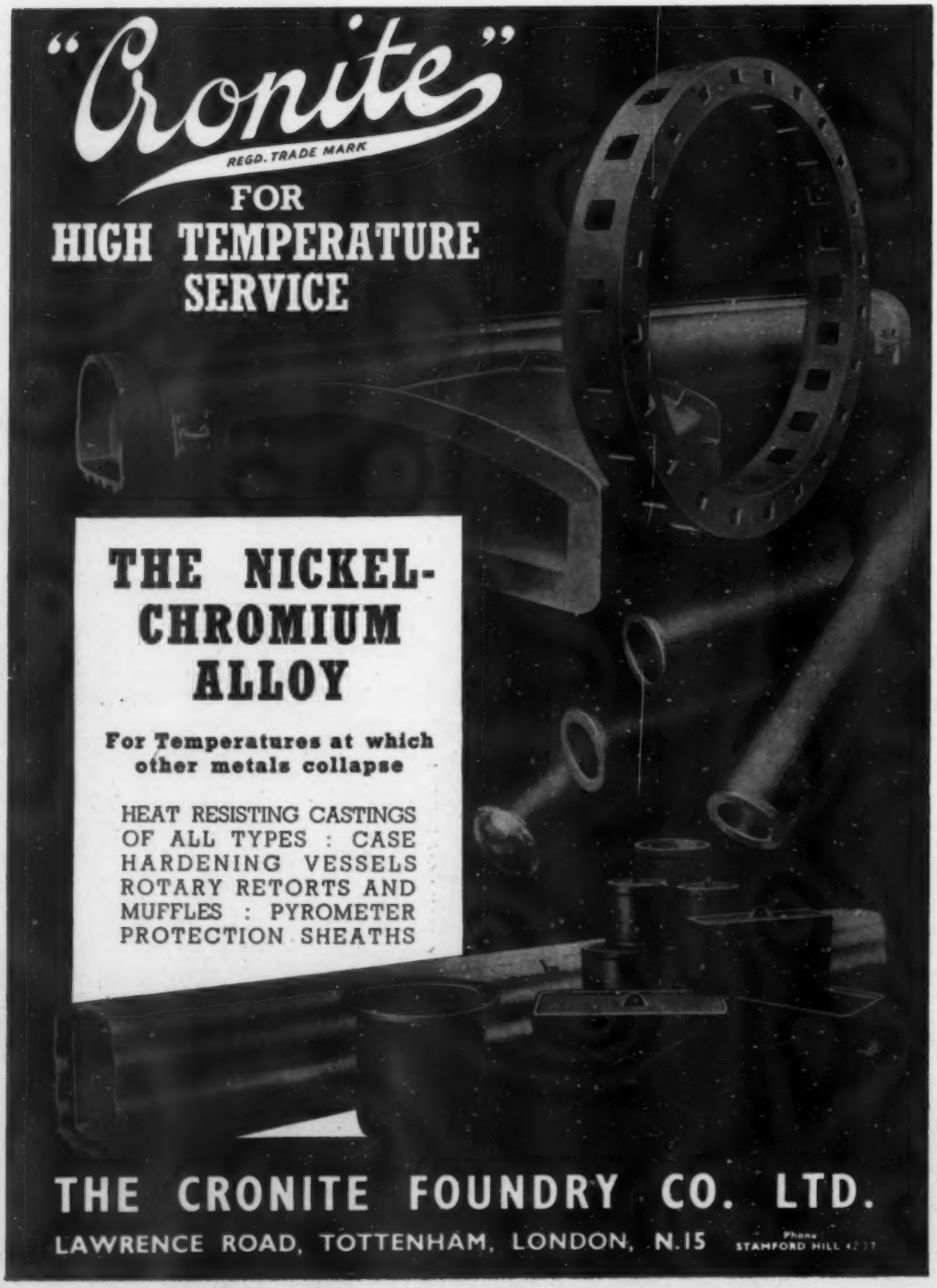
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METALLURGIA

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JULY, 1946 VOL. XXXIV No. 201

Castings and Weldings

THE question whether to make a casting or fabricate the component by welding frequently arises. It has been asked many times for 20 years or more, and will continue to be asked for many years to come. In the early days building up a shape by welding wrought material was regarded mainly as an emergency measure, and not intended as a new method for displacing castings. Following the last war, however, so much improvement in the technique of welding had been achieved that the tendency to use weldings instead of castings increased, partly due to the fact that the 1914-18 war disorganised the foundry industry, particularly iron founding; but also partly due to the lack of efficient organisation of the foundry industry at that time. The position has been repeated as a result of the recent war. Although to-day the industry is much better organised, the demands made upon it during the war have seriously disorganised its structure, which is causing many difficulties in effecting a change from war to peace conditions. The change is being rendered more difficult by the pressing need for castings to initiate reconstruction. Cast iron, for instance, is a basic material for so many products, and the demand for castings so overwhelming that it will be no surprise to learn that weldings in place of castings is on the increase.

The serious supply shortage of heavy iron castings led to the selection of the above title and subject for the Edward Williams Lecture presented at the recent annual general meeting of the Institute of British Foundrymen by Sir Claude D. Gibb, C.B.E., F.R.S. In the preparation of this lecture Sir Claude thought of laying down some broad rules with the object of guiding the designer in deciding whether a welding or a casting would best satisfy requirements in a particular instance. So many variables can affect the decision, however, that he ultimately abandoned the attempt. He came to the conclusion that manufacturing facilities available in a particular factory may make the decision for a designer, and over-rule technical considerations. The number required of a particular component may make the decision. The stresses applied in service—whether tensile or compression—may decide. The importance of weight, the need for a particular shape, these and more influence the choice. Indeed, the more carefully the subject is considered, the more difficult it is to define rules which would be applicable in every case.

Enthusiasts there are for each method of producing a component, and technical or economic considerations are not always considered without bias in one direction or the other. Shortly after the 1914-18 war the enthusiasm for weldings reached such a high point that the end of iron foundries was prophesied. The welding enthusiasts claimed that there was nothing produced as a casting which could not be made better as a welding. They made many mistakes, particularly in their claims to effect considerable saving in weight. It was argued that as the ultimate tensile strength of cast iron was some 12 tons/sq. in., and mild steel 28–32 tons/sq. in., the welding could be made 12/28ths the weight of the casting. Less than half the weight was the common claim. With certain components doubtless this claim could be substantiated, particularly in those few cases where a tensile stress had to be sustained and the welding could be designed to permit the steel bar or plate to carry the primary tensile stress without loading the actual weld. In such components as bedplates, however, it was found that, for equal satisfaction in service, the weight of the weldings was little, if any, less than castings, and the cost by no means in favour of weldings.

It is now more generally realised that both castings and weldings are needed in industry, and while there can never be any well-defined demarcation line to determine when components should be made as castings or as weldings, the greater the progress in one or the other method will cause overlapping, to the disadvantage of the fabricated product or the foundry industry, whichever is not progressive enough.

There are few in industry who do not appreciate the importance of iron castings. Cast iron is a versatile and valuable material which has innumerable applications. Apart from its high compression strength coupled with good tensile strength, as Sir Claude points out, cast iron has very high internal damping capacity, and is much less resilient than steel. Thus, when carrying a vibrating load, there is less likelihood of vibrations being transmitted through cast-iron than steel. Design modifications may make the steel entirely suitable, but there remains the fundamental vibration absorbing property of cast iron. Then, it must be remembered that the difference between the yield point and the ultimate tensile strength of cast iron is negligible, whereas, with mild steel it is considerable, excepting at the welds, which may be, and frequently are, the weakest parts of a fabricated structure. Cast iron, therefore, provides that degree of rigidity so necessary as the basic component of a machine or engineering structure.

It is the question of labour, however, which is causing the iron foundries most concern. During the war this industry suffered more than most industries, because it was a war of steel, rather than of cast iron, and the iron foundries were denuded of skilled labour either for the Forces or for other war operations considered to be more important than the production of castings. In order to cope with the work, iron foundries were asked to do mechanisation, which was increased to permit the employment of unskilled labour. To a large extent, female labour was used. Now that the demand for iron castings for peacetime activities has greatly increased, the skilled labour, of which the industry is so much in need, is not returning to the foundries as quickly as

they can be absorbed. Indeed, many have found more congenial or more lucrative employment in other industries, and some have become welders.

On the fabrication side the position is quite the reverse. The need for welders was so great during the war that special courses were instituted to train both male and female labour to satisfy the demand; thus, not only is ample labour available, so that fabricating by welding can contribute to reconstruction, but the tools and equipment are to hand; suitable jigs are all that is required in the change from war to peace activities, and, of course, these would need to be made for any new design of structure. Thus, it can be said that many engineering firms are favourably placed to fabricate structures by welding and since the delivery of castings, under present conditions, is a matter for conjecture rather than of fact, it is not surprising that weldings are coming into greater use instead of castings.

That weldings will make considerable headway there can be no doubt, the extent will depend upon the energy and progressiveness of the foundry industry. There is much to do to effect greater operating control in foundries and in applying the results of research to improve the properties of castings. The same applies to weldings. We do not yet know enough about them. The economic development of weldings will probably cause the production of structures very different from castings and the structures indicated by Sir Claude, composed of welding and casting, shows that drastic design modifications can overcome difficulties in particularly awkward components. He instanced an exhaust end of a turbine casing as a composite of casting and welding, not by any means the easiest job to produce as a casting.

Despite the inroads of weldings into what has been regarded as the activities of the foundry, conditions favourable to the foundry will bring about an adjustment when the choice of castings or weldings will be determined by service and economic considerations; but it is as well to remember that a welding or a casting is rarely an end in itself, and in considering the economic values of each the complete work and its service possibilities must be fully explored.

The British Cast Iron Research Association

Annual Visiting Day

THE above Association held its third annual open day for members at its headquarters, Alvechurch, near Birmingham, on July 10th. In view of the pressure under which members are working to meet urgent demands for castings, the attendance of 80–90 could be regarded as reasonable, particularly when associated with the interest of the visitors in the equipment installed in the laboratories and the work in progress.

This Association is very much alive and is doing valuable work in the improvement of cast iron products. It is noteworthy that, since July, 1945, fifteen confidential research reports have been issued to members in the Association's *Journal* and other reports have been given in its *Bulletin*. In addition, since January, 1945, some forty reports have been prepared for the consideration of Committees and Sub-Committees. As the post-war programme gains momentum the number is expected to increase. The paper presented to the recent Conference of the Institute of British Foundrymen on residual elements in whiteheart malleable, referred to elsewhere in this issue, is additional to the figures given above.

Enquiries from architects, industrial engineers, surveyors, builders and others interested in the building uses of cast iron are dealt with in the building uses department. The application to the ironfounding industry of the results of research is the work of the development department. A large part of the work of this department, however, is concerned with enquiries and problems submitted by members. Visits are arranged to investigate and assist members with actual problems which they are experiencing in their works and to apply the results of recent research to these problems. It should be noted that all enquiries from members are treated confidentially. This department has an engineering uses division which provides services for engineering foundries and users. The intelligence department collects information from all over the world and selects and arranges it for the use of staff and members.

Current Research Programme

The research department includes in its scope various service laboratories in which particular tests can be conducted for any department. These include microscopic examination, chemical analysis, machine shop, mechanical testing laboratory, experimental foundry, sands laboratory, etc, etc. The investigations carried out may be basic or fundamental in character and of interest to the whole industry, or they are devoted to the solution of particular problems. The current research programme includes the mechanism of graphite formation during solidification; the mechanism of graphite formation in the solid state during annealing, as in the whiteheart and blackheart process; the mechanism of decarburising cast iron as in the whiteheart malleable process; improved methods of analysis; completing work on the mechanical properties of acicular cast irons and general work on the study of high-duty cast irons; completing a monograph on the mechanical properties of cast iron in relation to design; the determination of gases in pig and cast iron by vacuum fusion methods; the study of the properties of ingot mould materials at elevated temperatures; fundamental work on the mechanism of corrosion of cast iron; work for the Engineering Castings Sub-Committee covering the properties of cast iron at elevated and sub-normal temperatures, internal stress, Meyer hardness indices, ladle additions, etc.; work for the Malleable Castings Sub-Committee on the life of annealing cans, production of test bars, the manganese-sulphur ratio, effect of fluctuating temperatures during annealing, inherent properties in raw materials for malleable iron, and gaseous atmosphere in annealing cans; work for the White and Chilled Castings Sub-Committee on factors affecting chill, chill-depth and type of mottle, and mechanical properties of white irons; work for the Light Castings Sub-Committee on heat-resisting irons, warping and cracking in light castings, and specifications for foundry materials; work for the Sands and Refractories Sub-Committee on the testing of foundry moulding and core sands, and studying the properties, especially plasticity, of moulding sands at elevated temperatures.

This Association has made many important contributions to the progress of the ironfounding industry and the above brief reference to current work indicates that further knowledge is being accumulated which will undoubtedly lead to improved quality of iron castings.

Attack of Various Superheated Steam Atmospheres upon Aluminium Bronze Alloys

By A. P. C. Hallowes,* B.Sc., A.I.M., and E. Voce,** Ph.D., M.Sc., F.I.M.

The action of superheated steam on complex aluminium bronzes has been studied as a result of some failures when used industrially in certain superheated steam atmospheres under service conditions involving stress. The investigation is described and the authors conclusions are given. This article is based on the second part of Report R.R.A. 527 of the British Non-Ferrous Metals Research Association, the first part of which is described in the June issue.

NOTE—*In the article published last month, the position of the tables gave rise to some confusion. Corrected reprints are available on request to the British Non-Ferrous Metals Research Association.*

CASES have been reported in which complex aluminium bronzes have suffered from attack when used industrially in certain super heated steam atmospheres and have cracked under severe service conditions involving stress.

In one of the cases reported, hollow truncated cones of aluminium bronze (D.T.D. 197; 10% Al, 5% Fe, 5% Ni), used as plungers in the manufacture of glass tumblers cracked when the exterior of the plunger was in contact with molten glass at a temperature of at least 800° C. and the interior walls cooled intermittently with canal water. Under these conditions cracking occurred on the water-cooled surface, the temperature of which was stated to be far in excess of 100° C., and the attack may have been due to superheated steam formed from water evaporated in the hollow interior of the plunger. In another instance a valve body, made from aluminium bronze containing iron and manganese, had shown very severe corrosion when used for controlling a superheated steam supply. It was reported that the boiler feed water had in this case been over-treated, and this would probably produce (by hydrolysis of magnesium and ferric chlorides) free hydrochloric acid in the steam.

To enable the first set of service conditions to be represented in the laboratory, a cylindrical specimen of D.T.D. 197 aluminium bronze, in which a hole had been bored, was heated to 600° C. and quenched internally with a fine jet of laboratory tap water. The heating and quenching treatments were alternated for approximately 250 cycles, and a micro-section was taken across the specimen. This, however, failed to show any signs of cracking or corrosion.

Later, extruded aluminium bronzes containing 10%, 11% and 12% of aluminium and the 10% alloy containing 2·8% of iron were tested in the steam of conductivity water in an all-glass apparatus, designed to avoid rubber connections and fitted with a refluxing system. In this way the steam and condensate were kept entirely free from sulphurous impurities, and under these conditions sections of the exposed surfaces of the alloys showed micrographically no evidence of attack after a period of 30 hours at 400° C. under atmospheric pressure.

Further trial experiments were made, in which the 10% and D.T.D. 197 aluminium bronzes were tested at 400° C. in the steam of tap water for 150 hours and in the steam mixed with approximately 1% of air for 125 hours. The 10%, 11% and 12% alloys and the 10% alloy containing 2·8% of iron were also exposed at 400° C. in the steam of extremely dilute sulphuric acid (N/10,000) for approximately 40 hours, but micro-sections prepared from the specimens showed that all materials were entirely free from cracks or attack.

Positive results were first obtained by contaminating the steam of tap water with a small amount of sulphur dioxide and passing the mixture over cylindrical specimens of aluminium bronze heated to a temperature of 400° C. in a silica tube furnace. The rate of steam flow was approximately 325 litres per hour, and the desired amount of sulphur dioxide was mixed with the steam by means of a flow-meter.†

The experiments on the 10% aluminium and D.T.D. 197 alloys were continued for 100 hours at 400° C., and the specimens were covered with a red, partially exfoliative scale at an early stage in the treatment. On the exterior of the 10% alloy a coppery coloured layer containing dark products could be seen, which was often separated from the unchanged core by a thin layer of α solid solution. The 2% and 5% alloys were then treated for 75 hours at 600° C., and micro-sections showed that the attack penetrated via the grain boundaries, but appeared to avoid the twin boundaries (Fig. 1). A thin layer of copper in which small recrystallising grains could be seen was present on the extreme exterior of these specimens (Fig. 2). The attack increased with the aluminium content in the case of the 2%, 5% and 10% Al alloys.

The layer of attack on the 10% alloy was approximately 0·009 in. thick after 100 hours treatment at 400° C., and in the case of the D.T.D. 197 material appeared to be similar in character, but penetrated rather less deeply.

Repetition of the experiments replacing the tap water with distilled water and contaminating the steam with a similar amount of sulphur dioxide produced corrosion of the same nature at a temperature of 400° C. The treatment was continued for 30 hours.

The reaction at a temperature of 800° C. in the same sulphur dioxide-steam mixture, though similar in nature, was more intensive than that obtained at 400° C. After 100 hours treatment a micro-section of the 10% alloy revealed a coppery coloured layer (Fig. 3) of approximately 0·045 in. average radial thickness on the exterior of the specimen, and separated from the unchanged core by an unbroken band of α solid solution, of much greater thickness than in the same material treated under similar conditions at 400° C. The products

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** Formerly of the British Non-ferrous Metals Research Association. Now with the Copper Development Association.

† The amount of sulphur dioxide passed into the steam was approximately 0·1% by volume, but later experiments showed that it was necessary for the concentration of the impurity in the apparatus to rise somewhat above this value (the rise being automatically effected by back-pressure from a condenser used for collection of the distillate) before the attack commenced and continued to any appreciable extent. The minimum amount of sulphur dioxide requisite to initiate attack was not precisely determined.

5% Aluminium Bronze after Treatment in Steam Containing a Small Amount of Sulphur Dioxide. Figs. 1 and 2.

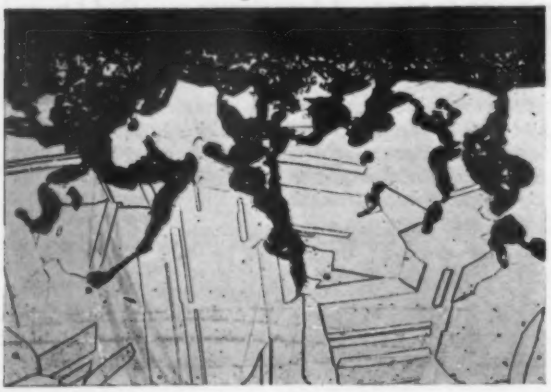

Fig. 1.—75 hours at 600° C. Etched with $NH_4OH + H_2O_2$ (× 150)
Intergranular penetration avoiding twin boundaries.

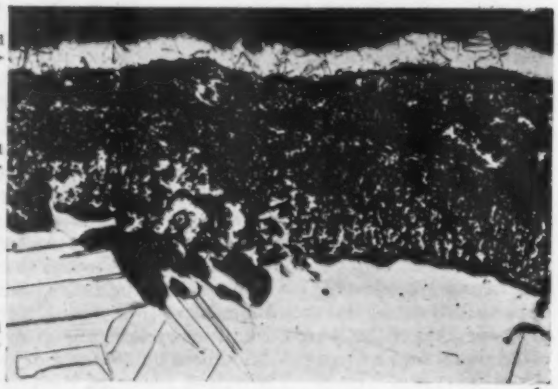

Fig. 2.—75 hours at 600° C. Etched with $NH_4OH + H_2O_2$ (× 150)
A different region of same specimen as Fig. 1, showing more extensive attack with thin layer of recrystallised copper on extreme exterior.

adjacent to this α layer appeared to consist largely of a bluish-grey constituent (Fig. 4) which failed under oblique illumination or polarised light to show the characteristic red colour of cuprous oxide.

One hundred hours treatment of the D.T.D. 197 alloy at 800° C. in the same atmosphere produced a similar, but rather thinner layer (Fig. 7), which was, however, much thicker than that obtained after similar treatment at 400° C. It was separated from the unchanged core by a discontinuous chain of particles of a bluish-grey colour similar to those observed in the 10% alloy. The attack showed a tendency to follow the distribution of the iron-rich particles.

Exposure of the 5% aluminium bronze to the same atmosphere for 100 hours at 800° C. gave a layer of the usual type. It was slightly thicker than that formed on this alloy at 600° C., but considerably thinner than upon either the 10% aluminium bronze or D.T.D. 197 at 800° C., indicating that increased aluminium content tended to increase the rate of attack.

10% Aluminium Bronze after Treatment in Steam Containing a Small Amount of Sulphur Dioxide. Figs. 3 and 4.

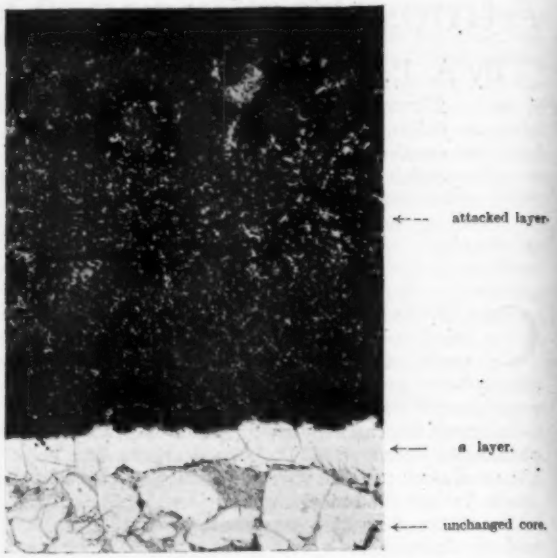

Fig. 3.—100 hours at 800° C. Etched with $NH_4OH + H_2O_2$ (× 50).
Extensive attack, with thick layer of α surrounding unchanged core.

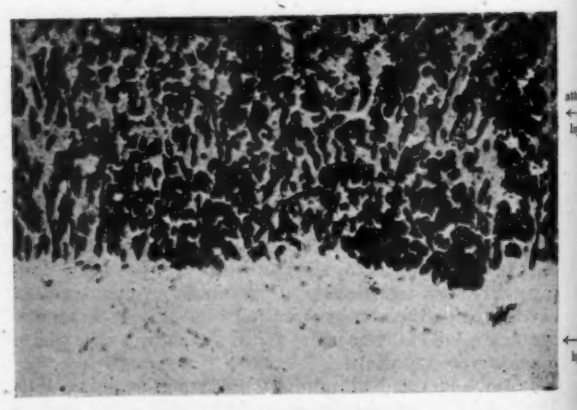

Fig. 4.—100 hours at 800° C. Unetched. (× 500).
Interface between advancing attack and α layer of Fig. 3. Bluish-grey and black products in coppery-coloured matrix.

The effect of replacing sulphur dioxide in the steam with a similar small amount of chlorine was examined by heating the 10% and D.T.D. 197 aluminium bronzes in this atmosphere for 70 hours at a temperature of 400° C. Micro-sections revealed that the surface corrosion was of a similar nature to that obtained when sulphur dioxide was used as the contaminating agent, and a layer coppery in colour and containing dark coloured products was again produced on the surface of the alloys.

To investigate any change of composition produced in the surface layers by the attack, the 10% alloy was heated for 170 hours at a temperature of 800° C. in

10% Aluminium Bronze after Treatment for 170 Hours at 800° C. in Steam Containing Approximately 1% of Sulphur Dioxide.

Fig. 5.—End of specimen showing disruption of attacked layer. (× 3).

Fig. 6.—Section of specimen showing thickness of attacked layer, the dark central area being the unchanged core (× 3).

steam containing sulphur dioxide increased in concentration to about 1%. The penetration after this treatment had reached a depth of approximately 0·1 in., and several pieces of the layer were cut from the specimen and examined.

Density determinations established a volume increase of 9·3%, which had caused the outer layers of the alloy to expand and rupture (Figs. 5 and 6) and their analysis was found to be :—

	%
Total copper	81·98
Alumina, present as such	15·33
Aluminium, present as metal	1·25
Oxygen (other than Al_2O_3)	0·72
Sulphur	0·05
Water soluble sulphur compounds	none detected.

The analysis showed that aluminium had been preferentially oxidised, and approximately 87% of the original aluminium content of the layer converted to alumina. It also suggested that aluminium was not lost from the alloy, since the Al : Cu ratio was 0·11 or the same, within the limits of experimental error, as in the original alloy. This was confirmed by collecting the condensed steam and some volatile matter which was deposited on the cooler portions of the reaction tube. The absence of aluminium salts in both was shown by chemical and spectrographic determinations.

The low sulphur content revealed by the analysis suggested that the part played by the sulphur dioxide was mainly catalytic. A sample of the layer adjacent to the unchanged core proved, however, to contain 0·1% of sulphur, and this indicated that the advancing line of attack may have consisted of alumina associated with sulphur compounds, representing the initial stages of the reaction.

The layer of α solid solution, enveloping the unchanged core, was in all probability formed as a result of the diffusion of aluminium in the direction of the outer layer impoverished of aluminium in solid solution.

The volume increase of the outer layers would become serious with extensive attack, and contribute to the failures reported under conditions of service. It may also cause sufficient porosity to permit diffusion of steam and gaseous impurity to the inner reaction face.

It is considered that the cracking of the plungers used in the manufacture of glass tumblers can be accounted for by the impure state of the cooling water and the steam produced from it. The reported case of the steam valve is probably to be explained by over-treatment of the feed water which, by hydrolysis, introduced free hydrochloric acid into the steam.

Comparative Tests on Brass and Tin Bronze

The experiments were finally extended to include 90/10 and 88/10/2 tin bronzes (which find wide application as valves and fittings for superheated steam supplies) and 60/40 brass. A study was made of the effects of a steam atmosphere containing sulphur dioxide upon specimens of these alloys in the sand cast condition.

After 100 hours at 800° C. in steam containing approximately 0·1% of sulphur dioxide the brass was covered with a light-coloured exfoliative scale, but was completely free from any penetrative attack. Micro-sections from the tin bronzes after the same treatment showed that they were also free from general penetration, though deep localised corrosion occurred either around porous areas in the casting or at points of contact between the specimen and its support. The penetration often reached a depth of as much as 0·05 in. at these points, and resembled that suffered uniformly over the surface of the aluminium bronzes under similar conditions, dark-

Fig. 7.—10% Aluminium bronze containing 5% iron and 5% nickel after treatment for 100 hours at 800° C. in steam containing a small amount of sulphur dioxide. Etched with $NH_4OH + H_2O_2$. (× 50).

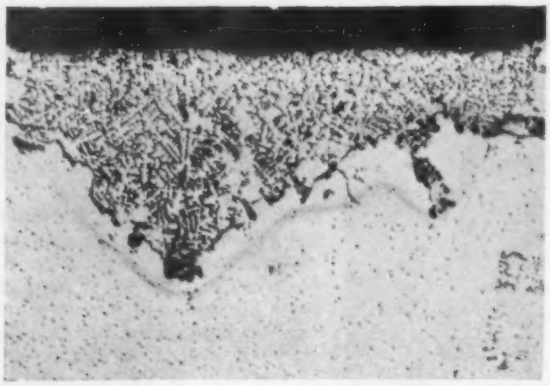

Fig. 8.—Cast 90/10 tin bronze after treatment for 100 hours at 800° C. in steam containing a small amount of sulphur dioxide. Unetched. Shows localised area of deep corrosion. (× 20).

coloured products being produced in a coppery-coloured matrix (Fig. 8). A tenfold increase in the concentration of sulphur dioxide caused similar localised penetration after 100 hours at 800° C., but failed to initiate general attack on these bronzes.

Microchemical examination of the products of the reaction taken both from the surface and the porous areas revealed the presence of crystals of stannic oxide coated with copper sulphide. Chloride, presumably derived from the tap water used to generate the steam, was also present and may have contributed to the reaction, though the experiment (see p. 119) upon aluminium bronze using the steam of distilled water, relatively free from chloride but contaminated with sulphur dioxide, suggests that the influence of chloride in the concentrations present in the steam of tap water was insignificant.

Summary and Conclusions

The action of superheated steam on aluminium bronze has been studied in connection with certain failures in service.

1. Pure steam was found to be harmless.

2. Prolonged exposure to an atmosphere of steam contaminated with a small amount of sulphur dioxide produced attack in the surface layers of aluminium bronzes at elevated temperatures. Preferential oxidation of aluminium to alumina occurred and the oxide remained embedded in a copper-rich matrix. A volume increase of more than 9%, leading to rupture of the surface layers, accompanied the reaction.

3. The attack was more severe at higher temperatures and with higher concentrations of contaminating agent in the steam; also when the aluminium content of the alloy was increased.

4. Attack of the same nature was obtained by using chlorine as the contaminating agent in the steam.

5. Specimens of 60 : 40 brass in the cast condition were not attacked after similar treatment, but, under certain conditions, cast tin bronze was subject to localised attack in which stannic oxide was found to be embedded in a coppery-coloured matrix.

The authors are indebted to the Director and Council of the British Non-ferrous Metals Research Association for permission to publish this paper.

Institution of Metallurgists

THAT the above Institution is vigorous and progressive is indicated by the fact that the first number of its *Bulletin* has been published. It will be remembered that the first ordinary general meeting was only held on November 28 last year, thus in a relatively short time the organisation has sufficiently advanced to render this achievement possible. As readers will be aware, this Institution was formed to promote, advance and co-ordinate the study of science of metallurgy in all respects; to promote in every possible way the interests of and to maintain and enhance the prestige of metallurgists and to encourage scientists whose main interests lie in metallurgy to become and designate themselves "Metallurgists."

The interest shown by metallurgists throughout the United Kingdom, as well as in the Dominions and countries of the British Empire has fully justified the belief of the founders that such an Institution would rapidly acquire the status, authority and representative character which will enable it to carry out its objects. In a Foreword by the President, Dr. Harold Moore, C.B.E., mention is made of the fact that in six months applications for admission indicated that a four-figure membership was well within sight. From the outset, however, the number of members has been limited, not by the number of applications, but by the capacity of the Membership Committee to give them the careful individual examination necessary to maintain the high standard set for admission. It is noteworthy that in the Report of the Council, published in the *Bulletin*, the number of members elected up to June 14, 1946, is 815, which must be very gratifying to the founders.

The following officers and members of Council have been elected :—

President: J. W. Jenkin, Ph.D., B.Sc., A.R.I.C. (Founder Fellow).

Vice-Presidents: G. W. Bailey, M.Sc. (Founder Fellow); Maurice Cook, D.Sc., Ph.D. (Founder Fellow); W. T. Griffiths, D.Sc., F.R.I.C., F.Inst.P. (Founder Fellow); R. A. Hacking, M.Sc. (Founder Fellow); R. Mather, B.Met. (Founder Fellow); and F. C. Thompson, D.Sc. (Founder Fellow).

Hon. Treasurer: C. J. Smithells, M.C., D.Sc. (Founder Fellow).

Members of Council Representing Associates: W. Barr, A.R.I.C. (Founder Fellow); W. E. Ballard, F.R.I.C. (Fellow); R. Griffiths, M.Sc. (Founder Fellow); H. O'Neill, D.Sc., M.Met. (Founder Fellow); W. K. B. Marshall, B.Eng. (Associate); and C. E. Homer, Ph.D. (Associate).

Members of Council representing Fellows: W. E. Alkins, D.Sc. (Founder Fellow); L. E. Benson, M.Sc. (Fellow); W. F. Brazener (Founder Fellow); E. W. Colbeck, M.A. (Founder Fellow); E. Gregory, Ph.D., M.Sc., F.R.I.C. (Founder Fellow); J. E. Hurst, D. Met. (Founder Fellow), N. P. Inglis, Ph.D., M.I.Mech.E. (Fellow); A. J. Murphy, M.Sc. (Founder Fellow); W. W. Stevenson, A.Met., F.R.I.C. (Fellow).

With such able and progressive officers and members of Council this new Institution will rapidly achieve its objects, and we believe our readers would wish us to express their good wishes for its success.

The Institute of British Foundrymen

Annual Conference held in Birmingham

It was fitting that the first post-war conference of this Institute should have been held in Birmingham, where it was founded in 1904, *and some evidence of the enthusiasm displayed is indicated by the fact that some* 690 *members and ladies took part.*

THE forty-third annual conference of the Institute of British Foundrymen, recently held in Birmingham, was admirably organised and well supported by members. It was the first occasion in six years that the normal arrangements associated with this Institute's conferences could be made, and it was particularly appropriate that the Birmingham Branch should have the privilege of organising the first post-war meeting, and to make arrangements for including ladies. Certainly the various committees responsible for organising the meeting, on the business, technical and social sides, are to be congratulated on the high degree of success achieved.

The conference opened with the President, Mr. John W. Gardom, in the chair, who, after extending a cordial welcome to members and visitors, made the following awards on behalf of the Council: The Oliver Stubbs Gold Medal to Mr. A. Phillips, for his outstanding work in connection with the foundry industry; the E. J. Fox Gold Medal to Sir Lawrence Bragg, F.R.S., presented by Mr. E. J. Fox; the British Foundry Medal and cheque to Mr. S. L. Fry. Diplomas for papers presented during the past session were awarded to Mr. N. C. Blythe, Mr. R. H. Brown, Mr. D. H. Young, Mr. A. V. Hill and Mr. J. Pearce, jointly, and Mr. T. Cook.

Election of Officers

Mr. D. H. Wood, who had been a vice-president for six years, was elected President for the ensuing year, and was invested with the chain of office; at his request, however, Mr. Gardom continued to conduct the business meeting. Mr. P. H. Wilson, O.B.E., and Mr. R. B. Templeton were elected senior and junior Vice-Presidents, respectively. The following members were elected Members of Council, to serve for two years: Mr. H. Bunting, Dr. A. B. Everest, Mr. C. Gresty, Mr. A. Phillips, and Mr. J. J. Sheehan.

Presidential Address

For his Presidential Address, Mr. Wood took as his subject, "Casting Reflections," in which he dealt with the early history of the Institute. Coming to the war years, he said that during the period 1940-44, British foundries produced more than 12,250,000 tons of iron castings, and 1,500,000 tons of steel castings, despite the difficulties of war and labour conditions, and of the official attitude towards the industry. Since the war, the President continued, the industry had switched over to peacetime work, particularly in the making of builders' castings. The recruitment of labour was still a major problem of the industry. Labour shortage was delaying production and the installation and use of mechanisation. There was a need for young men in the industry, and it was necessary to secure more rapid application in industry of the results of research. He made a hurried glance towards the future, and suggested some possibilities through questions, such as: Could not sand be conveyed under pressure direct from preparation plant to mould? Could there be a new version of Pluto conveying molten metal direct from the cupola to the mould? Could there not be more light and colour in foundries, improved dust and fume extraction, and was sand to remain the traditional moulding material? Might not some plastic material be developed for the purpose, which would be cleaner and more easily removed from castings.

Castings and Weldings

During the early part of the proceedings the Edward Williams' lecture was given by Sir Claude D. Gibb, C.B.E., F.R.S., who took for his subject "Castings and Weldings." Much in this lecture must provide food for thought, particularly for iron foundrymen, as there can be no doubt that the use of weldings instead of castings has grown, doubtless because of the war, and it is probable that it will continue to grow. As Sir Claude stated in his lecture: "There are to-day, and probably always will be, enthusiasts for weldings, and others for castings. Their enthusiasm frequently colours the analysis on which the decision is taken. The casting advocate claims that if there are several hundred off one article, and therefore pattern costs are spread over a large number of castings, the casting is not only much lower in cost, but more uniform in accuracy and quality. This may not be so if the welding is prepared and jigged with the same care as would apply to a machining operation. One thing the past war has taught us is the necessity to get away from the idea that welding is a means of producing a 'one off' article, or of repairing something broken in service. Welding has rapidly become a science, calling for high engineering and metallurgical skill. With the necessary control at all stages, weldings have a field in industry which is quite unchallengeable."

Technical Sessions

The three technical sessions held were well attended, and considerable discussion resulted. In addition, the many works visits arranged in connection with this conference proved both informative and interesting. It is not possible to deal with the whole of the papers presented, but one of the papers, presented by The Malleable (Metal Compositions) Sub-Committee of the Technical Advisory Panel to the Director for Iron Castings, Ministry of Supply, was of special interest because it deals with a problem likely to be experienced for some years to come, because of the presence of alloying elements in war metal scrap. Although it is primarily concerned with the influence of residual elements on the properties of malleable cast iron, the results of the work may have a wider significance, and brief reference to this paper and the subsequent discussion will be of interest. It should be noted that two further papers dealing with the annealing rate of gaseous

malleabilising and on the process of gaseous malleabilising with special reference to the use of steam, were discussed at the same time.

The Influence of Raw Material on the Properties of Whiteheart Malleable Cast Iron with Special Reference to the Influence of Residual Elements

THE investigations which form the subject of this Report, which was introduced at the meeting by Mr. J. W. Gardom, Mr. H. Morrogh, and Mr. A. E. Peace, were undertaken as a result of difficulties encountered in the production of tank track links in whiteheart malleable cast iron. These castings were subjected to an oil-hardening treatment after annealing. Frequent failure to meet the mechanical test requirements for the annealed and heat-treated components on the part of individual manufacturers, caused some concern by reason not only of the frequency but also the irregularity of these incidents. Some manufacturers were using cupola charges containing refined iron, whilst others used charges containing hematite and steel scrap, and differences had occurred in the properties of malleable castings produced from these two different types of mixtures.

In the production of whiteheart malleable castings, steel scrap is used either direct in the case of those manufacturers who use hematite and steel scrap, or indirectly in the case of those using mixtures of refined iron in the production of which steel scrap is used. This utilisation of steel scrap was regarded as a possible source of residual elements such as nickel, chromium, molybdenum, etc. Also, under the special wartime circumstances of shortage of hematite, refined iron makers had been requested to substitute ingot mould scrap as far as possible for virgin hematite. As it was known that much of this ingot mould scrap was burnt or badly "oxidised," it was also suspected that this may have been a possible cause of the mechanical test failures in castings from those foundries using refined irons.

In consequence of these difficulties, representations were made to the Technical Advisory Panel to the Directors for Iron Castings, resulting in the formation of the Malleable (Metal Composition) Sub-Committee with the following terms of reference :

(*a*) To explore the differences which have occurred in the properties of malleable castings produced from mixtures containing refined iron and those produced from mixtures containing hematite pig-iron and steel scrap.

(*b*) To investigate the influence of residual elements, which may be present in refined iron and hematite, used in the manufacture of malleable cast iron.

(*c*) To determine if, when burnt ingot mould scrap is used in the production of refined iron, this adversely affects the casting properties of the molten malleable iron, and has a deleterious influence on the properties of the annealed malleable castings.

As a result of the work done, the following general conclusions were drawn :

The analyses of the materials supplied for the investigation did not reveal any substantial differences between refined iron and hematite-base material in respect of their residual elements. In general, the refined base irons had higher carbon contents and lower sulphur contents. Whilst in general the phosphorus contents of the hematite irons were lower than those of the refined irons, their ranges of composition for this element overlapped.

Nickel in annealed whiteheart malleable cast iron can be present in amounts up to 0·3% without harmful effects.

Copper in amounts up to 0·7% (maximum studied) had no adverse influence on the strength of the larger sections ($\frac{5}{8}$ in.–$\frac{7}{8}$ in.), but tended to reduce the ductility of the small sections ($\frac{3}{8}$ in.).

In this investigation chromium in amounts up to 0·2% did not have any harmful effect on the mechanical properties of the annealed material. With a shorter annealing cycle, lower silicon contents and higher sulphur contents, the dangerous chromium content may be lower.

Molybdenum in amounts up to 0·2% (maximum studied) did not have any harmful effect on the mechanical properties. Molybdenum is a carbide-stabilising element, and, as such, its effects may be harmful in otherwise more difficult graphitised irons, or in irons subjected to a shorter annealing cycle. This effect may be more apparent with the simultaneous presence of chromium.

Additions of ferro-silicon-titanium and ferro-carbon-titanium did not have any marked influence on mechanical properties, but the graphitisation of the malleable iron, as indicated by its microstructure and the change of dimensions on annealing, was increased, as was also the tendency to mottle indicated by the thicker sections of the stepped test bar.

Tin had a feebly harmful effect, but in amounts up to about 0·1% in an otherwise good malleable iron the harmful effects of this element may not be appreciable.

Arsenic may be slightly harmful in amounts of the order of 0·1–0·2%, but the effect is not pronounced. In an otherwise good malleable iron, smaller amounts might be tolerated without noticeable effect.

Small additions of boron (as ferro-boron) of the order of 0·003–0·004%, may produce a marked increase in strength and graphitisation, but also tend to reduce ductility. It is not yet clear whether these effects are due to the alloyed boron, or to the deoxidising effect of the aluminium present in the ferro-boron used in these experiments. However, a melt to which equivalent amounts (0·001–0·005%) of aluminium were added did not show the effect. It was found that 0·04% boron completely inhibited graphitisation, and so it is probable that the critically dangerous amount of this element is considerably lower.

Phosphorus tended to reduce the ductility of small sections and the shock-resistance of all sections in the annealed condition. Its effects on other mechanical properties are not clear. Other factors being equal, the best general mechanical properties are most likely to be obtained with the lowest phosphorus contents. The phosphorus content may be as high as 0·07–0·08% for most purposes, and the difference between irons containing 0·05% and 0·07% phosphorus may only be detected statistically. It is suggested that the lowest phosphorus contents are desirable when good shock-resistance is required.

The carbide stabilising elements chromium and molybdenum tended to reduce the amount of expansion on annealing of the larger sections and to increase the amount of contraction in the small sections. Tin and

boron (in amounts greater than 0·1%) behave in a similar manner.

The graphitising elements nickel and copper tended to increase the amount of expansion on annealing of the larger sections and to reduce the amount of contraction of the small sections. Phosphorus behaved in a similar manner. It is possible that titanium and boron (in amounts of the order of 0·003%–0·004%) exert a similar graphitising effect, but this needs to be confirmed.

Heavy relative decarburisation, such as is experienced in $\frac{3}{16}$ in. sections, tended to obliterate the effects of the residual elements, except where they produced nothing in the original casting. This does not necessarily imply that the effects of the residual elements on the larger sections can be completely eliminated by an increase in the annealing time.

No "peeling" was encountered on any of the test bars used. It would appear that the residual elements did not favour the occurrence of the phenomenon under the conditions met with in this investigation.

No significant differences between refined iron and hematite steel melts were reproduced or recorded in the investigation.

The effect of treatment of melts with hydrogen is obscure, but such treatment may have a harmful effect on mechanical properties.

"Deoxidation" of refined iron melts with up to 0·02% aluminium, 0·02% calcium as calcium silicide, and up to 0·1% silicon as ferro-silicon-titanium, had no effect on the mechanical properties. The latter addition, while improving the annealability (graphitisation) of the metal, was liable to give free graphite in the unannealed castings, particularly in the larger sections.

In this investigation no evidence was obtained that melting in contact with oxidised material caused the production of inferior malleable iron.

The investigation surveyed, examined and indicated the limiting tolerable amounts of the most commonly-occurring residual elements. It is important to realise that the experiments used a hematite-base material, but the individual conclusions should be equally applicable to both hematite and refined iron charges used as melting stock in the whiteheart malleable iron industry.

No harmful effect was detected as the result of the use of oxidised scrap in the preparation of refined iron, on the annealability and mechanical properties of the resulting malleable iron.

Discussion

Mr. T. Makemson said he felt that the Committee had carried out all that was wanted of them, and even more. The Iron and Steel Control was more than satisfied with the work done by the Panel. Quite apart from the immediate object of the Committee, he believed it had provided a great deal of fundamental information which would be of assistance to every worker in the field of whiteheart malleable castings. The war made calls upon the malleable industry which it had never known before, not only in production—production for two years was 200,000 tons a week—something approaching double what it had known before, but also in the kind of jobs required. Technically, that needed something very much in advance of anything the industry had been called upon to do, either in blackheart or whiteheart. The industry responded nobly, and gave all that was required of it.

Dr. A. L. Norbury said that one of the chief conclusions he drew after studying some results, was that some factor was making the results very erratic, and was largely obscuring the effects of the added elements. The factor which was causing the irregular results was probably "mould conditions," due to the effects of varying amounts of mould gases entrapped in the metal. The Sub-Committee had realised at the outset that erratic results were being obtained, and had achieved somewhat better consistency by altering the design of the runners and feeders, but it had not been possible to investigate that aspect thoroughly since that would have delayed unduly the main investigation.

Another conclusion which he thought should be drawn was that the repeated blow Charpy Test results in the paper had given more useful information than the strength and malleability tests. Consequently, that type of test should be further investigated with the object of developing a standard shock-resistance test for malleable iron.

The important question of how much phosphorus could be tolerated in whiteheart was, unfortunately, left rather undecided by the results in the paper. The effect of boron was noteworthy and very interesting, and in that connection it should be remembered that molten cast iron could dissolve metallic boron from borax in slag.

Mr. H. Morrogh said that the actual results plotted were average values of several such results. He agreed that the Charpy Test was a useful one, and could profitably be developed further with whiteheart malleable, as it lent itself to such tests with unmachined test pieces.

Mr. J. Roxburgh congratulated the Committee on the excellence of their work and the conclusions drawn as the result of it. From the practical standpoint, he thought there might have been some suspicion about the soundness of the test bars from the manner in which they were fed and run. The diagram showed the two bars joined together by two feeders with the runner in between. Some time ago in collaboration with the laboratory, he found with certain test bars very small skinpools occurring which resulted in the lowering of the physical condition of some of the bars. He wondered whether the Committee had any evidence of that in their examination of the bars. One of the conclusions of the Committee was that copper in amounts up to 0·7% had no adverse influence on the strength of the larger sections. He wondered whether the Committee had come to any conclusion regarding any percentage of copper which might impart beneficial influence. As to the amount of phosphorus in whiteheart, he believed it was fairly well known that whiteheart was manufactured with phosphorus above 0·1%.

Mr. Peace said that Mr. Roxburgh was quite right in pointing out the question of the test bars and the possible graphitising defects. In his experience there was a cold lap which usually occurred in the centre of the bar where the two streams of metal came together. That might lead to some exclusion of gas. With the particular bars which were the subject of test prior to the tests being made, samples were cast with wooden patterns. After investigation the method was adopted which was likely to produce the soundest bar, but, even so, there were bars discovered in the heavy sizes which showed some micro-porosity. In no case was a cold lap discovered. There might be cases of porosity.

Mr. Morrogh said he believed he interpreted the results correctly by saying that they found no evidence that copper had any beneficial effect upon the mechanical properties. With regard to the remaining point, that whiteheart malleable with above 1% phosphorus had been produced, the Committee had been aware of that, but the investigation was begun to deal with certain problems of a very special critical nature, and they were anxious to uncover the slightest bad effect which any element might have. He had no doubt that many castings with more than 1% of phosphorus would be satisfactory for service expected from them.

Mr. I. Jenkins added his congratulations to the Sub-Committee for an excellent paper which was of very great practical value, but also raised one or two fundamental issues. Although the practical nature had been stressed, he was personally disappointed that the Committee had not ventured their views on certain of the more fundamental aspects of the investigation, because he was sure that it had many other views on the matters. There was the effect of titanium and boron, for example. The effect of titanium in cast iron apparently was quite contrary to its effect in steel. In steel it was recognised as a stabiliser, but with cast iron it had the reverse effect. Was it really the effect of the titanium melting, or the effect of some product of titanium after its addition to the melt, which accounted for titanium having the property of dissolving its own oxide?

Mr. Morrogh said Mr. Jenkins had put forward some ideas which were worthy of perhaps deeper thinking than was possible just at that moment. It was well known that the graphitising influence was not due to the influence of alloyed titanium. In that respect, titanium behaved somewhat similar in manner to aluminium. It was almost certainly related to the conversion of the oxide content of the iron into another form; a form containing aluminia or titania or zirconia. He would not like to be drawn too deeply into that theory. It was interesting; but it seemed to involve the solution of titania. Titania probably completely lost its identity in the metal, and therefore the theory seemed hardly tenable. He had not met the theory put forward with regard to boron, though it was an interesting one.

Mr. H. G. Hall said that certain rather lengthy conclusions had been drawn. He believed that the malleable industry was on the verge of a new era, and that technical advancement onwards would be very rapid. He was particularly anxious that pig-iron manufacturers should not allow the conclusions in the investigation to enable them to step up or at any rate, to retain the residual elements in any quantity.

Winding up the discussion, the Chairman, Mr. P. H. Wilson, senior, said: "As a pig-iron producer, and having made many thousands of tons of refined iron, all I can say is, Why don't you publish your specifications? We will work to them."

Sheffield Metallurgical Association

Formation of Special Groups

For some time there has been a desire by some members of the above Association for the formation of special groups. The Development Committee of the Association, after some preliminary inquiries, was satisfied that representation made to the Council for the formation of a group particularly interested in methods of chemical analysis involving physical principles, and another in refractory materials, justified meetings being arranged. The meetings were held on June 18 and 21, respectively, at which it was decided to proceed with both schemes provisionally. A temporary committee was formed in both cases to consider the constitution of the group, and to consider the conditions whereby members of other organisations might be invited to co-operate, to draw up a provisional programme for the autumn, and to arrange an inaugural meeting.

In the case of the physico-chemical analysis group, the Chairman of this committee is Mr. E. J. Vaughan of the Bragg Laboratories, and the Secretary, Mr. Eborall of Messrs. Geo. Senior & Sons, Ltd. For the refractories group, the Chairman is Dr. White of the University of Sheffield, and the Secretary, Mr. Howie of the United Steel Cos., Ltd., Stocksbridge. Each group desires the co-operation of other institutions and individuals, and all in the Sheffield area who are interested are asked to seek further information from the Secretary of the group concerned.

At a recent meeting the following gentlemen were elected and have accepted Honorary Life Membership of the Association: Dr. C. H. Desch, F.R.S.; Mr. Stuart C. Goodwin; Mr. W. H. Higginbotham; and Mr. Gerald Steel.

The Institute of Welding

Mr. A. Dyson, M.Inst.C.E., Joint Managing Director of Horseley Bridge and Thomas Piggott, Ltd., has been elected President of the Institute of Welding, and Mr. J. L. Adam, C.B.E., M.I.N.A., Chief Surveyor, British Corporation Register of Shipping and Aircraft, Vice-President for the year 1946/7.

At the recent annual general meeting of this Institute the retiring President, Mr. W. W. Watt, Managing Director of the British Oxygen Company, Ltd., gave an account of his recent tour of Australia and New Zealand and of the development of welding in those countries, when responding to a vote of thanks for his past services to the Institute.

Royal Aeronautical Society

At the recent Council meeting of the above Society, Sir Frederick Handley Page was elected President of the Society (his second term of office) for 1946-1947. The following were elected Vice-Presidents for the same period: Sir Oliver Simmonds, Dr. H. Roxbee Cox, and Mr. N. E. Rowe.

Sintering Plant for Dorman Long

We are advised that Messrs. Dorman Long & Co., Ltd., have placed an order for a complete sintering unit with Messrs. Huntington, Heberlein & Co., Ltd., 114, Cromwell Road, London, S.W. 7, who are licencees for the well-known Dwight Lloyd continuous sintering machines.

Erratum

"*The Importance of Ultimate Extension as an Engineering Property of Material.*"

On pp. 83 and 84 of our issue for June, 1946, the references D.T.D. specifications as 281 and 285, should read 285 and 289, respectively.

Refining Noble Metals, their Alloying and Fabrication

The Recent Exhibition by Johnson, Matthey & Co. Ltd.

The working of silver, gold and metals of the platinum group and their application to industry have been the primary occupation of Johnson, Matthey and Co. Ltd. during the past hundred years. During the past twenty-five years, however, their range of products has been enlarged to include specialised base metal alloys designed to meet unusual requirements and those who were privileged to visit the Exhibition, recently held at Dorland Hall, London, by this Company, would appreciate the unique range of its activities. Here it is only possible to describe, very briefly, some of these activities.

TO those fortunate enough to visit the Johnson, Matthey exhibition at Dorland Hall last month, the remarkable range of activities would probably prove to be the most striking reaction. Although primarily concerned with the industrial applications of the precious metals—silver, gold and metals of the platinum group—for many years the unique experience of this Company has been applied to increase the range of their products to include specialised base metal alloys designed for special purposes. In addition, the possession of the plant and equipment essential for the manipulation of these metals and alloys to exacting specifications enables the Company to fabricate many of the more common base metals to unusual degrees of precision in the form of fine wires, small seamless tubes, thin strip, accurately machined parts and multiple assemblies.

Products in the various fields of activity were grouped in the various divisions embracing industrial, jewellery and allied trades, dental surgical and optical, chemical, and ceramic divisions. Of these, the industrial division is the most extensive, incorporating six main sections according to the needs of industries served, such as, chemical plant and apparatus, instruments, electronics, switchgear, resistance welding, and low-temperature brazing, brief reference to which is made in the following notes.

Chemical Plant and Apparatus.—The use of silver and platinum in chemical process plant depends chiefly upon their resistance to attack by a wide range of active reagents, their useful mechanical properties and, in the case of silver, its excellent thermal conductivity. Fine silver is highly resistant to attack by organic acids, by halogens, by dilute solutions of certain mineral acids, and by alkalies and salts. Silver plant is therefore used in the manufacture of pharmaceutical preparations, organic acids, food-stuffs and photographic materials. When operating conditions are extremely severe, as in handling corrosive liquids and gases under pressure at high temperatures, in melting optical glasses, and in electrochemical processes, platinum, usually employed as a lining or sheath, has advantages over all other materials.

In this section the exhibits included silver and silver-lined process equipment and accessories, such as stills, jacketed pans, condensers, liners agitators, tubes, cocks and pipe fittings. While applications of platinum were shown in the form of electrodes, agitators, sulphuric acid sprays, bursting discs, glass melting crucibles and jets for fibre glass production. Platinum has always been indispensable for the construction of laboratory apparatus and examples in this field included crucibles, dishes, micro-chemical apparatus, and electrodes for electrochemical analysis. Several types of platinum wound high temperature laboratory furnaces were also shown. In addition silver, platinum and palladium catalysts were shown in a variety of forms.

Instruments.—For the manufacture of pressure and temperature measuring instruments, the products exhibited included Bourdon tubes in phosphor bronze and beryllium copper, strip in the same alloys for the production of diaphragms and springs, accurately drawn capillary tubing, platinum; rhodium-platinum and other thermocouples, and platinum for resistance thermometers. Materials for electrical instruments were represented by fine resistance wire, hair spring strip in phosphor-bronze, beryllium-copper and cadmium-copper, galvanometer suspension strips in phosphor bronze, fine silver, gold-covered silver and gold suspension alloy, fine aluminium tube for pointers, vitrified enamel scales on glass, and precision machined repetition parts in brass, beryllium-copper, silver and other materials. Whilst for use in optical instruments there were graticules, consisting of gold lines and scales fired on to glass to give a permanent design, and optical filters, also produced by firing on to glass gold films to give specific visual transmissions and heat reflection.

Electronics.—This section included resistance wires in nickel-chromium and copper-nickel, together with typical potentiometers, attenuators and fixed resistors. A range of nickel cathode and other tubes for valves and cathode-ray tubes was also on view, together with tubes and pressed parts in a special alloy for use as glass to metal seals. The precision drawing of both round and rectangular tube in copper, beryllium-copper, silver-clad copper and other materials, in conjunction with the electrodeposition of silver and rhodium, enables this Company to provide radio manufacturers with components for wave guides, resonant cavities, coaxial lines and their associated sliding bridges and contacts. Other products included terminal insulators in ceramic materials, metallised for soldering to metal containers, a range of silvered mica condenser plates, and fine fuse wire for telecommunication purposes.

Switchgear.—A very comprehensive range of electrical contacts were shown in this section, comprising 45 metals and alloys, including silver, gold, platinum, palladium, iridium, rhodium, copper, tungsten, molybdenum and their many alloys and combinations, each in a number of different forms. Contact springs were

also shown, in phosphor-bronze, nickel-silver and beryllium-copper, as well as examples of contact sub-assemblies produced to users' designs. Allied to the use of silver as an electrical contact is its very wide use in high rupturing capacity fuses, based upon its freedom from oxidation, its high electrical and thermal conductivities and its low specific heat. Examples of silver fuse elements produced from wire, strip and pressed parts were included in this section.

Resistance Welding.—Successful resistance welding is now an accurate and controlled process in which pieces of metal are assembled rapidly and economically by welds of consistent strength. Much of this achievement is due to the provision of hard, high conductivity alloys for use as electrodes and dies. The range of alloys for this purpose has been developed by the Company over many years, each material being suitable for a particular duty or range of applications. This section included standard spot-welding electrodes, light, medium and heavy duty electrode holders and adaptors, seam welding wheels and blanks, and Elkonite inserts and Elkonite faced dies for projection, butt and flash welding.

Low Temperature Brazing.—Silver brazing alloys were featured in this section which showed the range of forms in which they are available, such as strip, wire, rod, foil, wire, rings, washers and other forms suitable for pre-placing assemblies to be brazed by mechanical methods. Demonstrations in the use of these silver brazing alloys added to the interest.

Some Recent Appointments

MR. JOHN G. ALLEN, Commercial Manager, and Mr. JOHN E. MERCER, Foundry Manager, of the steel foundries of Head, Wrightson & Co. Ltd., have been appointed Directors of the Stockton Steel Foundry Co. Ltd., a subsidiary company.

MR. J. HEPBURN, formerly Commercial Director of Davidson's Sirocco Engineering Works, Belfast, has been appointed Managing Director of Musgraves & Co. Ltd., St. Ann's Ironworks, Belfast. Mr. Hepburn is to undertake the reorganisation of Musgraves, which has been established over ninety years.

DR. J. L. M. MORRISON has been appointed to the Chair of Mechanical Engineering in the University of Bristol, vacated by Professor Robertson, who is to occupy the Chair of Engineering.

MR. R. N. HEATON, A.C.A., has been appointed Secretary to the British Non-Ferrous Metals Federation, with effect from 1st July, 1946. It is announced that the Director and Staff of this Federation now occupy offices at 132, Hagley Road, Birmingham 16 ('Phone Edgbaston 3386-7-8), to which all correspondence should be sent.

DR. F. D. RICHARDSON has been appointed Head of the Chemistry Department of the British Iron and Steel Research Association. A graduate of University College, London, and one time Commonwealth Fellow at Princeton University, U.S.A., Dr. Richardson played an important part in the development of new weapons for the Royal Navy. In H.M.S. *Vernon* he was associated with the development of equipment for sweeping magnetic mines, and he played a leading part in the development of the "wiping" method of degaussing. He was then transferred to the Admiralty, where from 1942-46, he was Deputy Director of Miscellaneous Weapon Development.

DR. W. A. ARCHIBALD has been appointed head of the Refractories Section of the Chemistry Department of the British Iron and Steel Research Association. A graduate of Glasgow University, Dr. Archibald has previously carried out research on steel slag refractories problems at the Royal Technical College, Glasgow. For the past seven years he has been with Messrs. General Refractories, Ltd. (Glasgow). As their Technical Manager he established wide contacts with the iron and steel industry which should be particularly useful in forwarding the application of new refractories developed to solve the problems of the makers and users of iron and steel.

MR. PATRICK R. BROGAN has joined the staff of Messrs. Paterson, Simons, Ltd., agents for Hadfield's, Ltd., of Sheffield, in the Malayan Peninsula. Mr. Brogan has had a long experience at the East Hecla Works of Hadfield's, Ltd., where he has specialised in heavy dredger work. This appointment is in accordance with the policy of Hadfield's, Ltd. in supporting their overseas representatives with expert technical assistance.

MR. B. BLACKWELL GREEN has been appointed representative for Edgar Allen & Co. Ltd. in Sheffield and district in succession to Mr. Gilbert Hallam, who has retired after serving the company forty-three years. Mr. Green was formerly works manager of the company's Trackwork Department, but he will continue to be available for users of that department's products, so that his specialised knowledge can assist users outside his actual area of representation.

MR. E. E. PAYNE, until recently General Manager of the machinery department of Edgar Allen & Co. Ltd. and the British "Rema" Manufacturing Co. Ltd. at Sheffield, has been appointed London Manager of the Engineering Division of Edgar Allen's in succession to Mr. H. W. T. Mason, who is retiring in August next. Mr. Payne is a Director of British "Rema" Manufacturing Co. Ltd. and his work in London will include the supervision of their interests.

Dr. L. N. BRAMLEY has been appointed Electrical Engineer to the Plant Engineering Division of the British Iron and Steel Research Association, where he will be working on problems associated with electrical equipment of steelworks plant generally, including power follow-up gear such as is found in steelworks cranes.

Mr. W. H. HADLEY, M.A., F.R.I.C., has joined the organisation of Messrs. T. J. Priestman, Ltd., Birmingham, as chief chemist and technical adviser. Mr. Hadley, who has been on the staff of the British Aluminium Co., Ltd. for about twelve years, will, in his new position, be largely concerned with the technical control and maintenance of the quality of the T.J.P. aluminium alloys.

Mr. M. SEAMAN, M.Sc., A.M.I.Mech.E., A.M.I.E.E., and Mr. J. F. B. JACKSON, B.Sc., A.R.I.C., A.I.M., have been appointed to the Board of P.R. Jackson & Co., Ltd., Salford Works, Manchester, 5, a subsidiary of David Brown & Sons (Huddersfield), Ltd.

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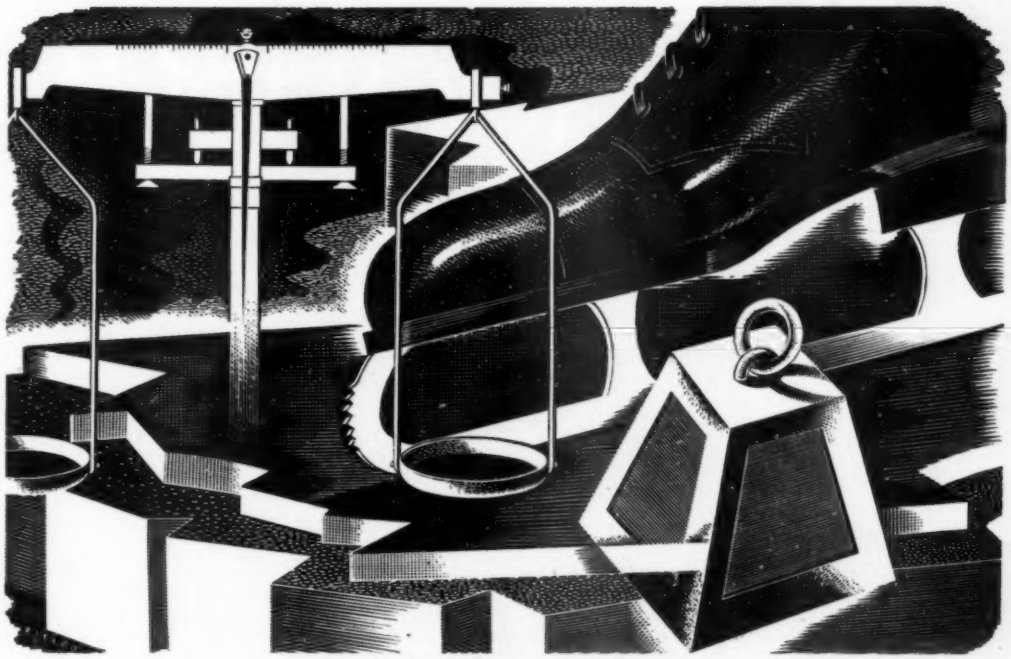

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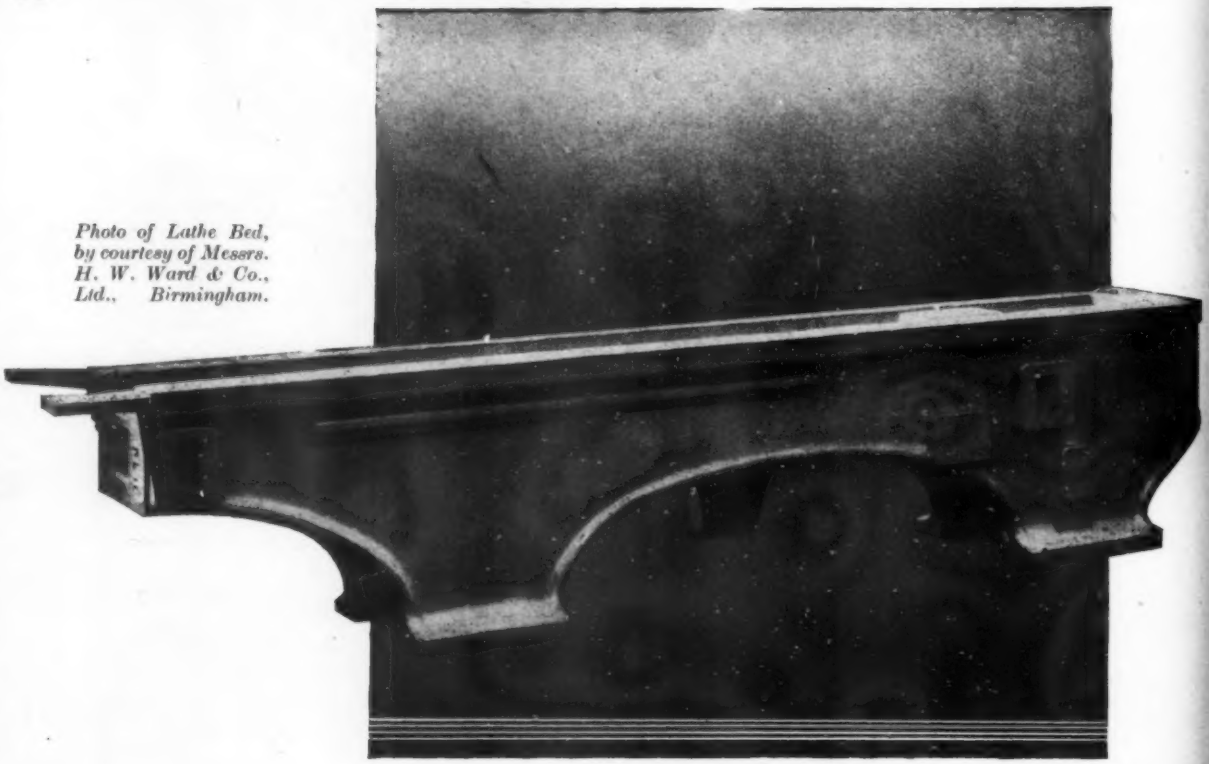

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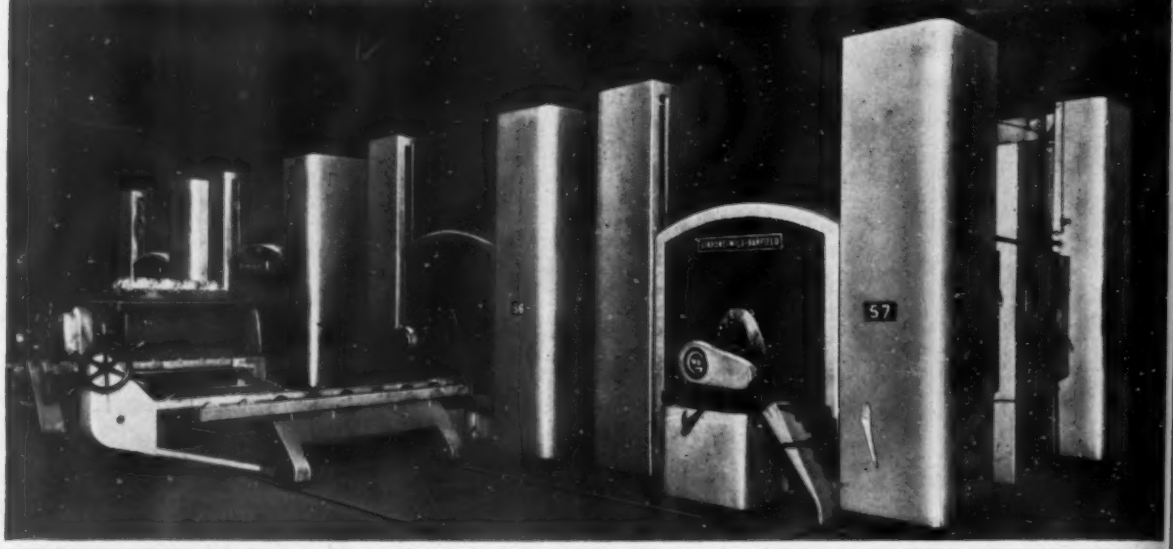

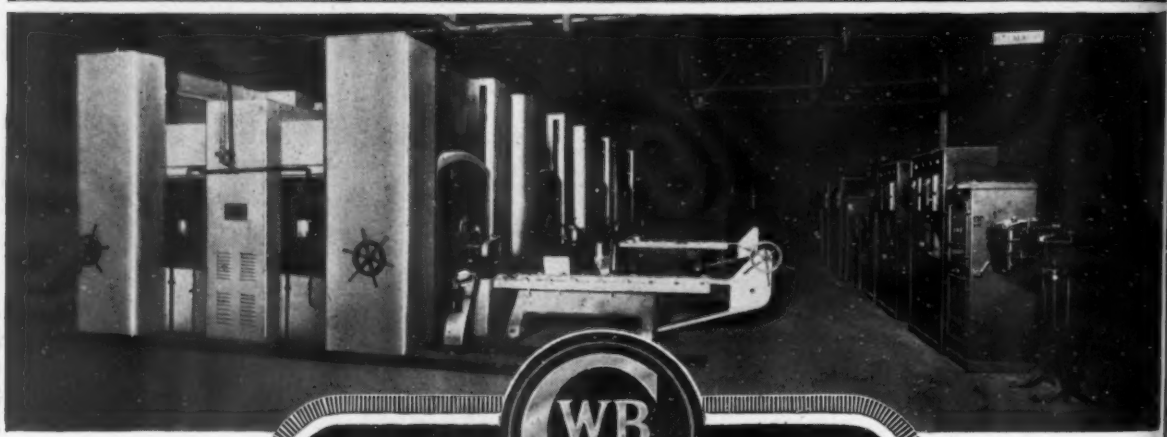

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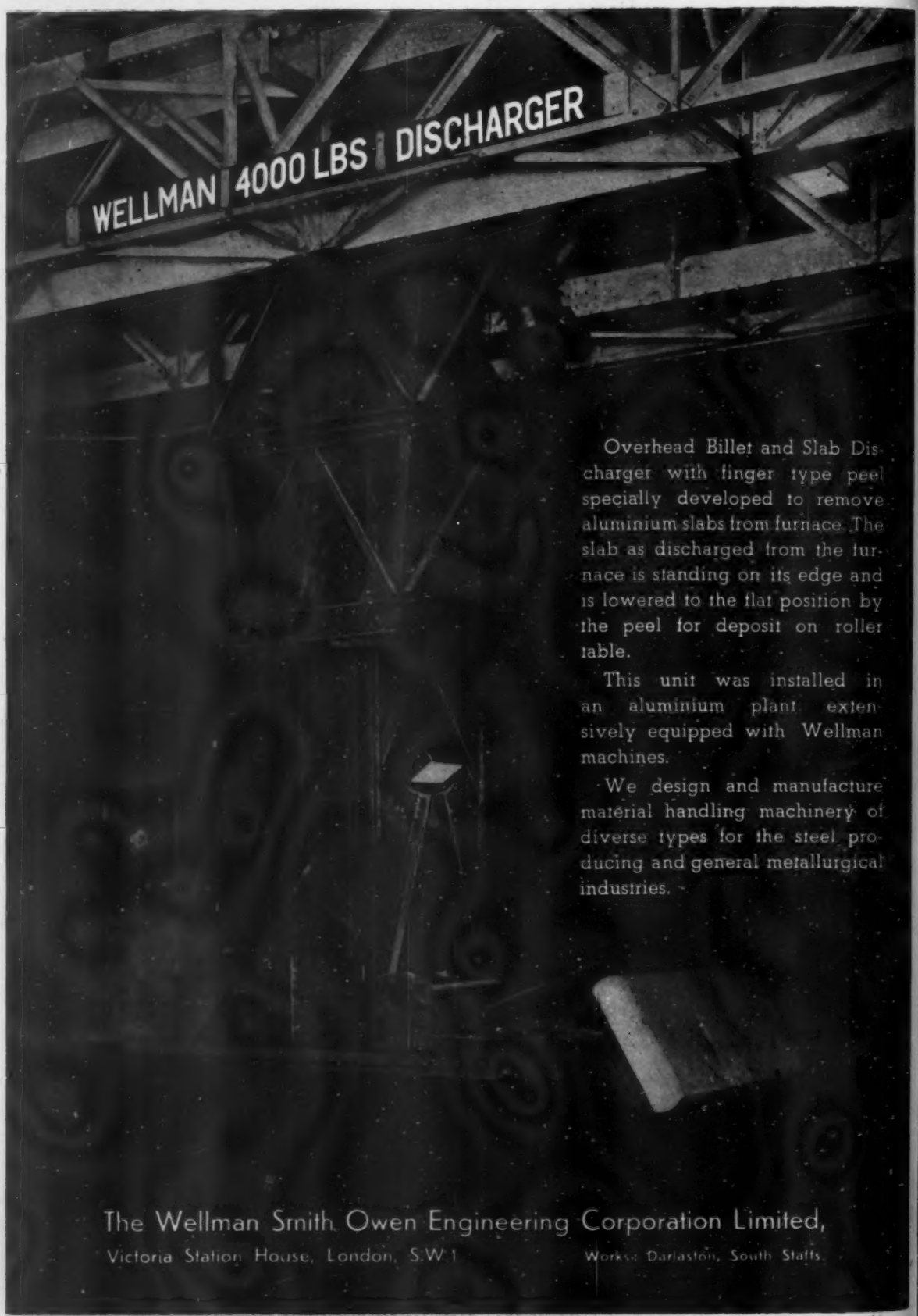

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A Small-Scale Creep-Testing Unit

By G. T. Harris, M.A., F.Inst.P.

(Research Dept., William Jessop & Sons, Ltd., Sheffield)

Much research work has been carried out to devise suitable apparatus and technique with the object of determining the strength of an alloy at elevated temperatures, particularly resistance to creep deformation. Further work has been carried out on short-time tests in order to accelerate the development of new heat-resisting alloys for gas turbines which resulted in the development of a small-scale creep-testing machine. This machine is described and some results are given showing the form of the strain/time curves obtained.

THE present state of the theory of metals is such that no certain prediction can be made of the strength of an alloy at elevated temperatures from its chemical composition. This applies particularly to the resistance of an alloy to creep deformation. It is, therefore, necessary in the development of new alloys to melt casts of the required analysis and to test these under creep conditions in order to obtain a true measure of the properties of the material at high temperatures.

The conventional creep-testing machine besides being bulky is expensive to install and maintain and, furthermore, the test-pieces require a comparatively large amount of material, demanding up to 1 ft. of 1-in. diameter bar for each test. The need, therefore, arose for a creep-testing unit which would be relatively inexpensive to construct and maintain; which would occupy the minimum of space, and which would use a test-piece small enough to enable a preliminary survey of high-temperature properties to be made from a cast of metal weighing only 10–20 lb. A further need for such a unit arose when tests were required from parts which had been in service and especially when the supply of material was limited. It has, for instance, been found possible to test samples prepared from gas turbine discs and blades and from aero-engine valves removed from crashed aircraft, with the unit about to be described.

Several types of testing machine have been built to meet the following requirements. The machine was to be as small as possible and was to take a test-piece requiring not more than 3 in. of $\frac{1}{4}$-in. diameter bar. Provision was to be made for stressing up to 30 tons/sq. in. and strain measurements were to be sufficiently accurate to enable creep rates down to 10^{-5} or 10^{-6} per hour to be measured.

The final test-piece has a gauge length of 1 in. and a cross-sectional area of $\frac{1}{40}$ sq. in., corresponding to a diameter of 0·1785 in. The loading capacity of the machine is then $\frac{3}{4}$ ton, which is too high for dead loading in view of the inconvenience of applying the load and the high stress that would be imposed on the framework of the laboratory building by a battery of such machines. The load is, therefore, applied through a 10 to 1 lever with individual weights ranging down to the equivalent of a $\frac{1}{4}$ ton/sq. in. No jockey weight is used.

Description of Unit

The general construction of the machine can be seen from

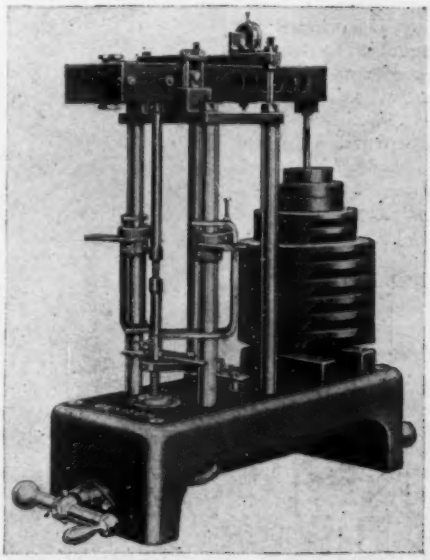
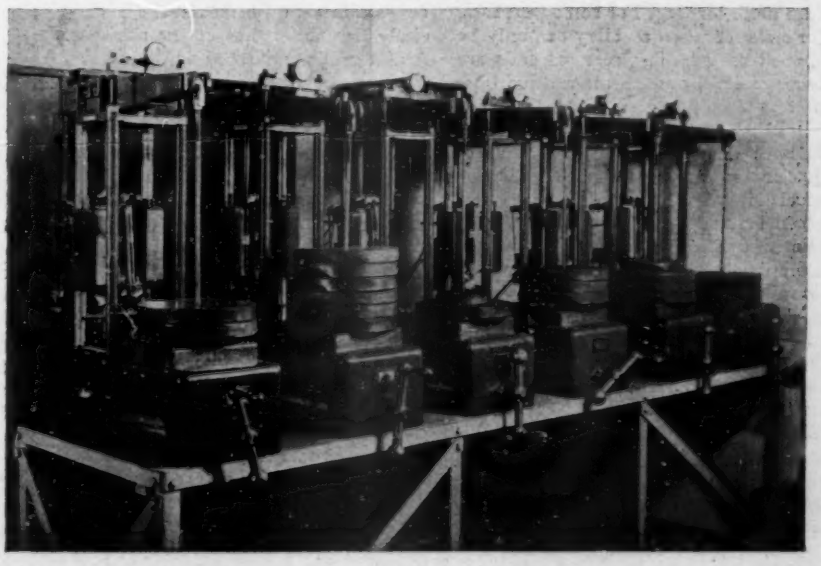


Fig. 1—Small-scale creep-testing unit frame.

Fig. 2.—Group of small-scale creep-testing units.



Figs. 1 and 2, and it will be noted that to keep the overall height of the machine small the lever is placed above the test-piece. The pull is transferred from the lever through universal joints to a heat-resisting steel shackle approximately 10 in. long. A similar shackle is attached by a pin-joint to a screwed bar which is raised or lowered by the rotation of a bevel gear carrying a square-threaded nut. The gear is rotated by a shaft carrying handles at both the front and back of the machine.

Two types of shackle are employed. For most materials the test-pieces have screwed ends with ¾-in. dia. British Standard Fine threads. These screw into the shackles which

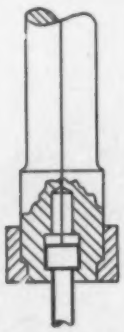


Fig. 3.—Shackle for headed test-piece

are very carefully machined to ensure axial loading. For some purposes, for instance when an alloy can only be shaped by grinding, screwed ends are inconvenient, and an alternative type of test-piece has a simple cylindrical-headed end. This test-piece is gripped in split shackles which are held together by a non-austenitic stainless-steel ring, which, having a lower coefficient of thermal expansion than the austenitic steel shackle, exerts a clamping force when the temperature is raised. (See Fig. 3).

The furnace consists of a stainless-steel tube approximately 7 in. in length by 1¼-in. in diameter, on which is wound a Nichrome winding spaced so as to obtain an even temperature distribution over the 1-in. parallel portion of the test-piece. As the shackles have considerable thermal mass the winding turns are close together at the two ends with few turns over the central portion. The tube is contained in a furnace casing about 5 in. in external diameter, supported by brackets from the main columns of the machine. A power input of approximately 150 watts maintains the test-piece temperature at about 650° C., but temperatures up to 900° C. can be obtained.

At first, no attempt was made to fit an extensometer measuring directly on the specimen as this entails considerable complication and in many cases is not found necessary. A

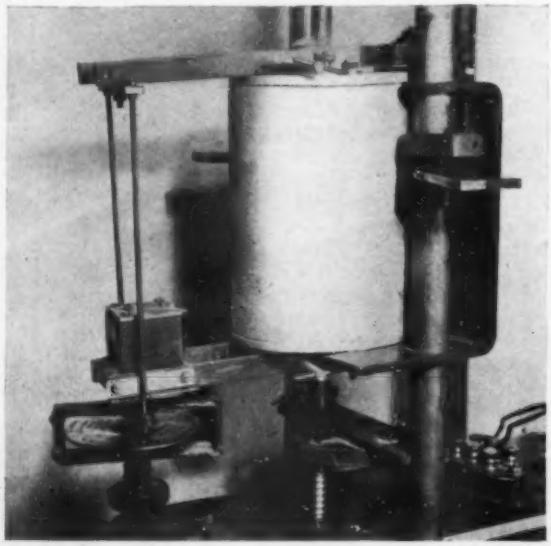


Fig. 4.—Expansion-type temperature controller.

hardened steel ball is mounted on the lever arm in the knife-edge line at such a position that its movement is precisely five times that of the shackles. A dial gauge calibrated in half thousandths of an inch is mounted above this ball, the dial-gauge plunger being fitted with a flat anvil in place of the usual ball-end, so that the magnification is fixed by the original construction of the beam. In locations subject to heavy vibration, stability of the dial gauge has been attained by mounting the whole creep-testing unit on four rubber bungs approximately 1-in. in diameter.

Accuracy

It will be seen that one division of the dial gauge corresponds to a strain of 10^{-4} or 0·01%, and by estimation, readings can be made to 10^{-5} strain or 0·001%. It was found that more sensitive strain measurements were not usually significant in view of the effect of furnace and room-temperature variations. For instance, a variation in test-piece temperature by 1° C. will alter the dial gauge reading by approximately half a thousandth of an inch (*i.e.*, one division), whilst a temperature variation of 1° C. in the frame changes the reading by approximately one and a half thousandths of an inch. It is evident that control of the frame temperature is important for accurate readings, but in practice the errors introduced are found to be small due to the fact that the room temperature is generally under control.

It is desirable to control the test-piece temperature to within about ± 1° C. Several methods have been tried. A simple and inexpensive control employing an energy regulator of commercial manufacture (compensating for mains voltage fluctuations by their effect on a bimetallic relay) gave regulation within about ± 2° C. but was subject to occasional instability, causing temperature changes of several degrees. Many of these units have, therefore, been operated using a resistance thermometer controller as described by Prosser[1]. The thermometer of platinum wire is wound non-inductively either over the furnace winding or on a silica tube adjacent to the test-piece. This control gives a temperature variation of about ± ¾° C. A further electronic method in use on a batch of these units incorporates a thermocouple, potentiometer, galvanometer, photo-cell and gas-filled relay, and gives control of about ± ¼° C., which is much closer than is normally required. Most of the units in use, however, employ a simple and inexpensive instrument giving regulation to about ± ½° C. In this method, the expansion of the furnace tube relative to an external structure of special low-expansion alloys is arranged to operate a micro-switch in the furnace circuit. The regulation obtained is very satisfactory and, by the use of special alloys of expansion characteristics intermediate between those of the 36% nickel steels and ordinary steels, the control is almost independent of variations in the ambient temperature. A photograph of one of these controllers is shown in Fig. 4. Advantages of these controllers are their low cost, good control, simplicity with little to go wrong and that they leave the whole of the furnace tube free for test-piece and thermocouples.

The test-piece temperature is measured at two points by two

1. Prosser, *Engineering*, 1939, 148, 95–96.

platinum/platinum-rhodium thermocouples in thermal contact with the test-piece and shielded from radiation by stainless steel caps. The necessity of this latter precaution was pointed out by Bailey[2], because serious error can be caused by direct radiation at temperatures of the order of 500° C. and above.

Creep Data

Although the unit was designed primarily for sorting tests to enable alloys to be arranged rapidly in order of creep resistance and to obtain comparative data on the effect of variation in composition and manufacturing methods, it is obviously important to know that the results obtained are comparable with those obtained on conventional large creep units. The results of one test carried out to confirm this are given in Table I below and also in Fig. 5, which give data on three tests cut from one bar of a chromium-silicon steel. Two of the tests, A and B, were carried out on two different types of large creep units (one of the single-lever and one of the double-lever jockey weight type), and the third, C, on a small-scale creep unit of the type described above.

TABLE I

Test	Type of Unit	Strain at 300 hours: per cent.	Creep Rate 250/450 hours per hour
A	Large	1·60	$3{\cdot}1 \times 10^{-6}$
B	Large	1·46	$3{\cdot}0 \times 10^{-6}$
C	Small	1·84	$3{\cdot}0 \times 10^{-6}$

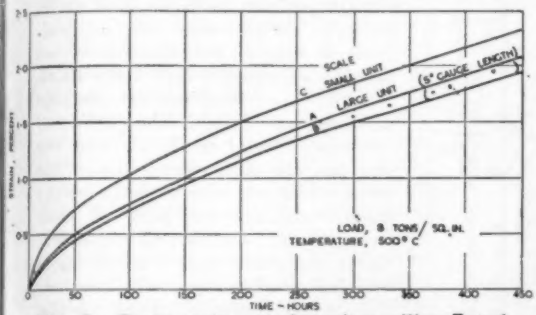


Fig. 5.—Creep tests on chromium-silicon steel.

It will be seen that the agreement in creep rates on the three tests is very close, being in fact probably accidentally good in view of the variation known to exist in creep rates. The actual strain values at a given time normally show a wider spread, being at such a relatively short time much more affected by the initial strain-hardening of the steel and by such factors as accuracy of alignment of test-piece and shackles. The small unit shows an apparently larger initial creep due to the fact that the strain measurement includes a certain amount of "bedding down" movement in the shackles.

TABLE II
CREEP TESTS ON CARBON STEELS
STRESS 8 TONS/SQ. IN. TEMPERATURE, 450° C.

Test	Steel	Type of Unit	Creep Rate Chord 24–48 hours
D	A	Large	$9{\cdot}0 \times 10^{-6}$
E	A	Small	$9{\cdot}6 \times 10^{-6}$
F	A	Small	$9{\cdot}6 \times 10^{-6}$
G	B	Large	31×10^{-6}
H	B	Small	29×10^{-6}

Further tests carried out to check the suitability of the machine for short-time check tests are quoted below in Table II and in Fig. 6. They refer to tests carried out on standard and small creep units with the object of determining the rate of creep between 24 and 48 hours under given conditions of stress and temperature, as laid down in British Standard Specification Number 1271, 1945.

It will be seen that even at this comparatively early stage in the creep curve the rates of creep measured in the two types of units are in good agreement. Experience has shown that the unit is satisfactory for the measurement, over such short periods, of creep rates down to 5×10^{-6} per hour, and over longer periods down to 2×10^{-6} per hour.

Newer Developments

Later developments of the machine included an automatic strain/time recorder and an extensometer for more refined measurements. The recorder is shown in Fig. 7 and consists of a drum rotated once in seven days by a synchronous clock motor, with a pen

[2] Bailey and Roberts, *Proc. Inst. Mech. Eng.*, 1932, 122, 209.

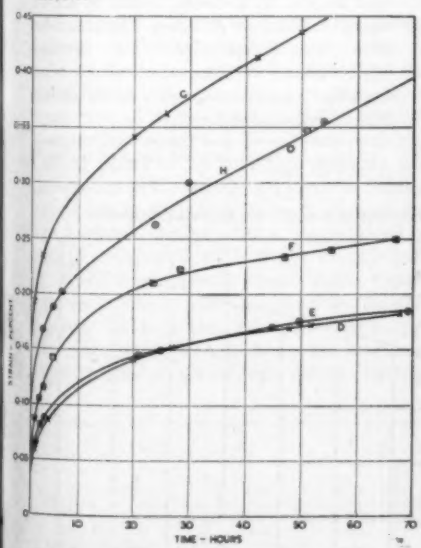


Fig. 6.—Short-time creep tests comparing large and small specimens.

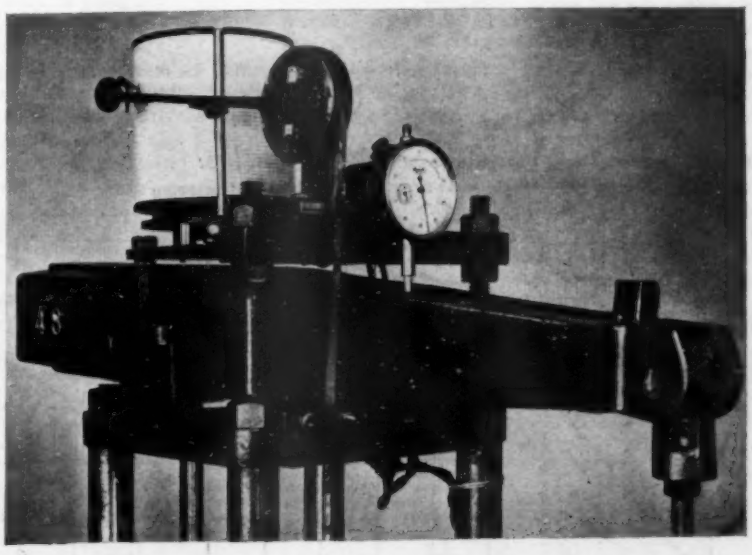


Fig. 7.—Autographic creep-strain recorder.

moved by the beam, and giving a magnification of test-piece movement of approximately forty. It has been found useful in showing the form of the strain/time curve particularly in the later stages of creep, when insufficient dial gauge readings are available, particularly during week-ends.

For the measurement of creep rates of the order of 10^{-6} per hour and less, and to determine the initial plastic strain, it was found necessary to use an extensometer attached to the specimen and not to the shackles or loading arm. A suitable instrument is shown in the photograph Fig. 8 and in the diagram Fig. 9. The upper shackle G is drilled to take the rod D whose lower end is turned down and has a fine rounded end which presses on the bottom of a hole drilled in the top-threaded portion of the test-piece A. At the lower end of the test-piece is machined a collar, the lower gauge-point, giving an effective gauge lengthof 1 in. Two legs B and C made from split tubes transfer the movement of opposite sides of the collar outside the furnace. A lever E so shaped as to have its action points on the same plane is sprung with H, H′ against the tube B, J against the rod D, and K against the flat anvil of a dial gauge F attached to the other leg C. The dial gauge is calibrated in 0·0001 in. and hence one division corresponds to a strain of 5×10^{-5} strain or 0·005%. The elastic strain for, say, 10 tons/sq. in. with a steel of Young's modulus (at the testing temperature) of 10,000 tons/sq. in., corresponds to 20 ten-thousandths of an inch, so that even with the usual errors of such gauges the modulus may be obtained with sufficient accuracy for many purposes. For higher sensitivity the dial gauge can be replaced by a rhomb carrying a mirror. Proof stress curves carried out using an instrument of this type with dial gauge are given in Fig. 10, and a creep curve in Fig. 11.

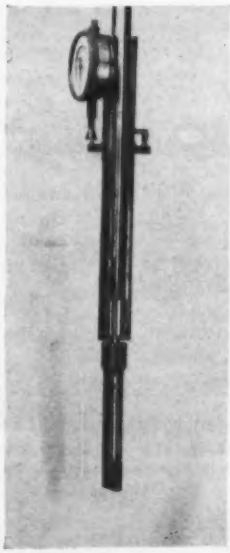


Fig. 8.—Small-scale creep extensometer.

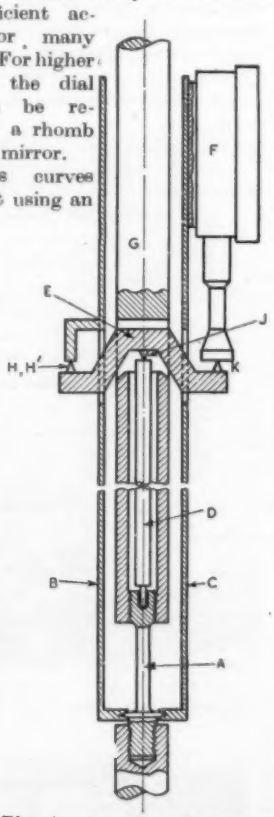


Fig. 9.—Small-scale creep extensometer.

Apart from the suitability for such short-time tests as described above, it has been found that the use of a battery of these machines with tests of the order of 300/500 hours has accelerated the development of the new heat-resisting alloys widely employed in gas turbines[3] by enabling the necessary initial sorting tests to be carried out with the least possible expenditure of skilled labour, money and test-piece material. The much more elaborate creep tests on large units are then only carried out on alloys which have passed beyond the initial experimental stage and on which accurate design data are required. Such a simple long-time test has definite advantages over tests which depend on relaxation or on the measurement of the initial portion of the creep curve when all but the simplest alloys are concerned. The more advanced creep-resisting alloys usually depend for part of their strength on complex precipitation-hardening phenomena, of which a false impression can be obtained if adequate time is not available for the reaction to develop along its normal course.

Acknowledgments

The author's thanks are due to Mr. D. A. Oliver, M.Sc., F.Inst.P., Director of Research, Wm. Jessop & Sons, Ltd., Sheffield, in whose laboratories these machines were developed, and to Mr. A. Lenton, A.M.I.M.E., of Samuel Denison & Sons, Ltd., Leeds, who was responsible for certain instrument refinements and who supplied the photographs from which Fig. 1 was taken.

3. Oliver and Harris, *Iron and Steel*, Vol. 19, No. 7, p. 379, June, 1945.

Fig. 10.—Proof-stress test with small-scale extensometer.

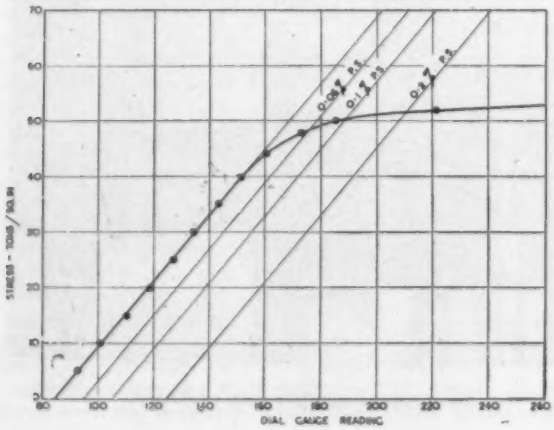
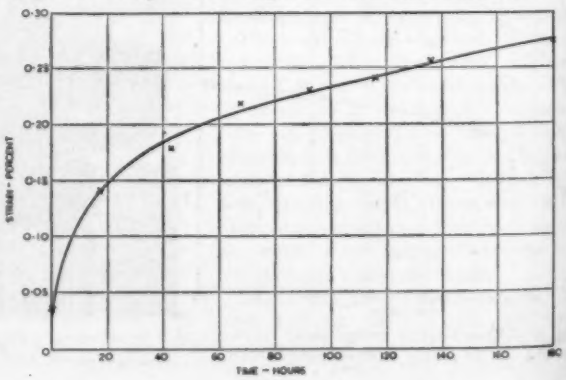


Fig. 11.—Creep curve using small-scale extensometer.



Developments in Low Alloy Steels for Welding Construction

By A. J. K. Honeyman, B.Sc., F.I.M. and J. Erskine, B.Sc., A.I.M.

In view of the remarkable progress in welded construction, a new demand has arisen for constructional steels of high tensile strength combined with satisfactory welding properties. The British Institute of Welding has laid down a DPH of 350 developed in a fillet weld as being the maximum for a weldable steel. This limits the carbon content of the steels to be used for this purpose. Here a range of 0·14–0·20% carbon is given and in order to obtain a yield strength of 25 tons/sq. in. the effect of alloy additions is considered and particular attention is directed to a manganese-molybdenum steel which satisfactorily fulfils requirements.

THE progress made in automobile and aircraft engineering design has caused a constant demand for materials of higher and higher tensile strength. This led to the increasing development of high alloy steels which, in the quenched and tempered condition, give tensile strengths up to 100 tons/sq. in. without serious loss in ductility, and the measure of success attained by the use of these heat-treated alloy steels is indicated by the continued decrease in the weight/power ratio of internal combustion engines.

The rapid advance in automobile and aircraft engineering appears to have stimulated mechanical and structural engineers and they too began to seek steels of higher strength for plates and sectional material. Economic considerations, however, ruled out the alloy steels which were available and a much cheaper product was required. The demand was met by the introduction of steels of higher carbon content with manganese and sometimes a little chromium as the alloying elements. These steels are supplied in the "as rolled" or normalised condition since quenching and tempering is an expensive process. For example, B.S.548 provides for a steel with a tensile strength of 37–43 tons/sq. inch and a minimum yield point of 23 tons/sq. inch for plates and sectional material not exceeding 1¼ in. in thickness. The carbon content is specified to be 0·30% max. Table I shows a typical analysis and mechanical properties of this steel compared with those of mild steel.

Such high tensile steels have proved perfectly satisfactory when used in riveted construction, but recent years have witnessed remarkable progress in welded construction and a new demand has arisen for constructional steels of high tensile strength combined with satisfactory welding properties. This new requirement automatically set a limit to the carbon content of the steel as will be readily understood from the following brief discussion.

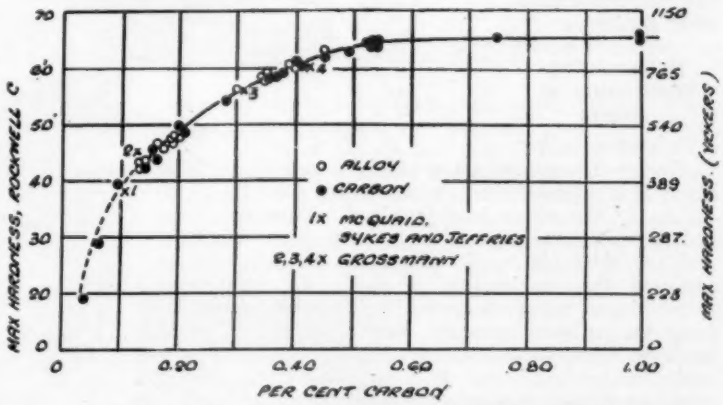


Fig. 1.—Maximum hardness V S carbon content. (J. L. Burns, T. L. Moore and R. S. Archer.)

Effect of Carbon Content

During welding, the steel adjacent to the arc undergoes a thermal cycle which involves a temperature gradient from the melting point of the steel down to a temperature of 200–300° C. over a very short distance. All conditions of heat-treatment may be met with in this small area, ranging from a drastic quench down to a mild softening. Consider the effect this has on a plain carbon steel as the carbon content is increased from 0·05–0·4%. Through this carbon range the hardness on quenching is increased from 223 DPH to 770 DPH.

Fig. 1 is a curve of the maximum hardness developed as the carbon content of steel is increased. It is also shown in the figure that this effect is independent of

TABLE I

Steel	Analysis					Mechanical Properties			
	C %	Si %	Mn %	Ni %	Cr %	Yield Point Tons/sq. in.	Max. Stress Tons/sq. in.	Elong % 8 in.	Red. of Area %
Mild (B8.15b)	0·20	0·05	0·60	—	—	15·2	29·8	24	52
H.T. (B8.548)	0·25	0·05	1·48	—	—	24·5	41·2	20	49
H.T. (Welding) (B8.968)	0·22	0·05	1·35	0·25	0·10	21·5	38·2	21	60
(STA5V6B)	0·19	0·05	1·10	0·30	0·15	19·5*	35·8	23	52

*0·2% Proof Stress

alloy additions. These do not increase the maximum hardness obtained but only facilitate its development. It should also be noted that in a plain carbon steel the maximum hardness is reached only under ideal quenching conditions but, as already indicated, the possibility of reaching the maximum hardness becomes a reality with alloy additions.

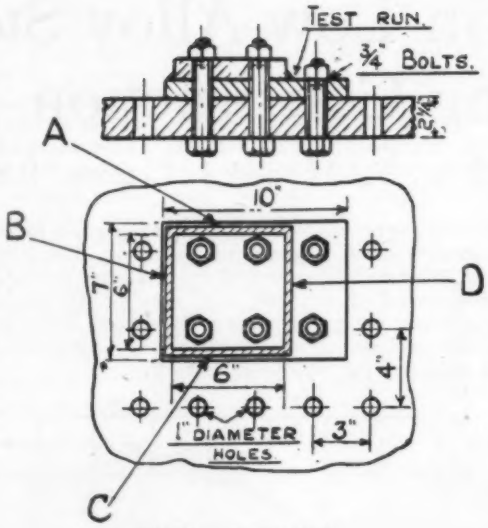


Fig. 2.— Reeve test.

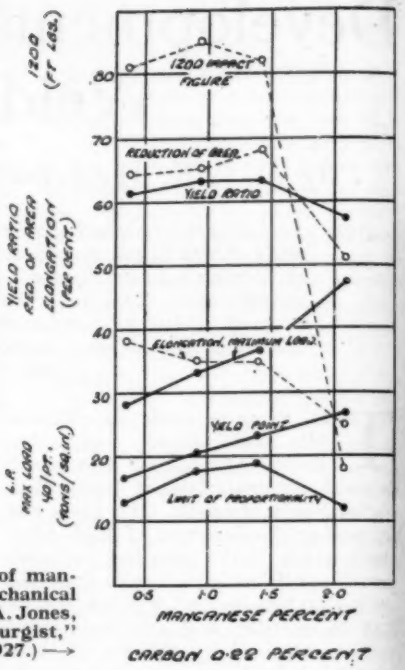


Fig. 3.—Effect of manganese on mechanical properties. (J. A. Jones, "The Metallurgist," Oct. 28, 1927.) →

Determining Weldability of Steels

Before proceeding to discuss the various types of high strength weldable steels it is advisable to consider briefly how the weldability of the steels is determined. Throughout the literature various types of tests have been described and considerable work done in an attempt to evaluate the weldability of steels, e.g., Tee bend tests, single run tests, patch weld tests, the Reeve test, etc. More recently the attempt has been made to correlate the welding characteristics of a steel with the Jominy Test[1]. It is not the purpose of this paper to discuss their relative merits, and, to compare the steels, the authors have employed only the Reeve test. This test is carried out under conditions more severe than is usually met with in ordinary welding practice, and can be satisfactorily applied to the study of pre-heating and post-heating. A brief description of the Reeve test is given hereunder, but for fuller details readers are referred to the original paper[2].

The test equipment consists of a heavy base plate to which are bolted two plates of the material under test, as shown in Fig. 2.

The test plate (Fig. 2) is fillet welded round sides A, B and C, and the whole allowed to cool. When cold, any slack bolts are tightened and the test run is made in side D, using the electrode and technique under test. When steels are being compared, the electrode and technique are kept constant. When cold, the welded block is removed and the test fillet sectioned and examined visually for any cracking. A diamond pyramid hardness survey is also carried out in the thermally disturbed zone of the parent plate. This hard zone occurs to a greater or less degree in every weld and constitutes a serious problem since the hardness is accompanied by brittleness which can result in serious failure. After extensive investigations, the British Institute of Welding laid down a diamond pyramid hardness number of 350 developed in a fillet weld of approximately 0·046 sq. in. area as being the maximum figure allowable for a steel to be considered weldable. This figure is considered by the authors to be satisfactory.

1 *Journal American Welding Society*, July, 1943, p. 278—S.
2 *Transactions Institute of Welding*, 1938, vol. 1, p. 13.

Effect of Alloy Additions

To consider now the effect of alloy additions; as already stated, when the alloy content is increased so also is the hardening capacity of the steel, until a stage is reached when straightforward air cooling is sufficient in itself to cause the material to develop its maximum hardness. It will thus be realised that the element which must first be carefully controlled is carbon and

TABLE II

Steel	C %	Si %	Mn %	Ni %	Cr %	Mo %	Other Elements	Yield Point (Min.) Tons/sq. in.	Max. Stress Tons/sq. in.
BS.968 ..	0·22 max.	0·35 max.	1·8 max.	0·5 max.	1·0 max.	—	Mn + Ni + Cr 2·0 max.	21·0	35/41
ST–A5V6B.	0·21 max.	0·05/0·35	0·8/1·20	0·4 max.	0·25 max.	0·15 max.	Mn + Ni + Cr + Mo 1·5 max.	18·5*	30/36
Si-Mn ..	0 23 max.	0·70 max.	0·7/1·0	—	—	—	—	21·5	33/38
2% Ni ..	0·20 max.	0·3 max.	0·40/0·80	2·0 min.	—	—	—	50% of M.S.	34/38
Yoloy ..	0·05/0·35	0·2 max.	0·30/1·0	1·5/2·0	—	—	Cu 0·85/1·10	23·5	30·4 min.
Corten ..	0·12 max.	0·25/0·75	0·20/0·50	0·65 max.	0·50/1·25	—	Cu 0·25/0·55, P 0·07/0·15	22·0	31·0 min.
Mn-Va ..	0·13/0·18	0·15/0·30	1·20/1·45	8—	—	—	V 0·10	24·5	35·5/42·5
Ni-Mo ..	0·18/0·25	0·3	0·7 max.	1·8/2·2	0·3 max.	0·2/0·3	—	29·0	40/48
C–Mo ..	0·13/0·25	0·15/0·30	0·40/0·60	—	—	0·50	—	18·8/25·4	32·0/35·5
UX1 ..	0·14/0·18	0·3 max.	1·0/1·4	0·5 max.	0·25 max.	0·20/0·30	—	25·0	35/41

* 0·2% Proof Stress.

Fig. 4—C 0.22%, Mn 1.4%, ½ in. thick plate, 8 SWG. electrodes. →

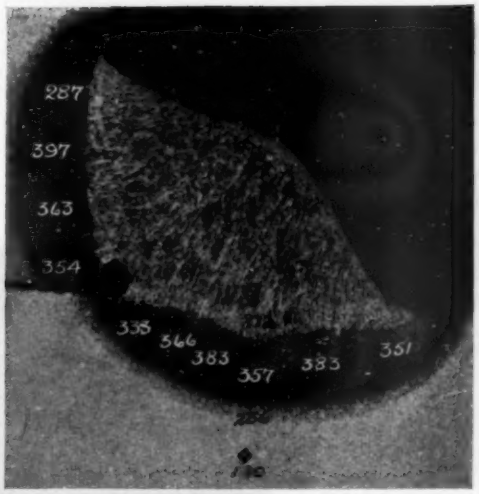
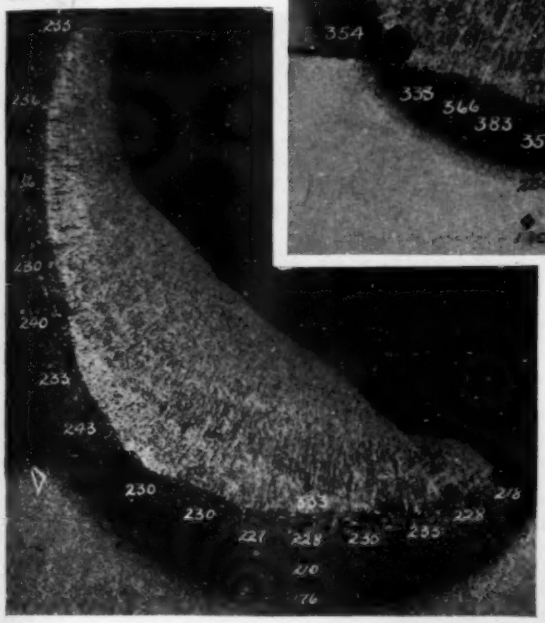


Fig 5.—C 0.22%, Mn 1.4%, ½ in. thick plate, $\frac{5}{32}$ in. dia. electrodes.



thereafter the mechanical properties developed by suitable alloy additions.

While it is recognised that almost any steel can be welded provided appropriate precautions and safeguards are taken, by weldability the structural engineer means the ability to be welded satisfactorily with normal working conditions and with ferritic as distinct from austenitic electrodes. On this basis the authors are of the opinion that the carbon content should be as low as possible having regard to the required mechanical properties, and in any case, should not exceed 0·20%. Since for a working range in the manufacture of steel, limits closer than ±0·03% are not practicable, the lower limit of carbon must be 0·14% and thus the range of carbon content which should be taken as the basis of a low alloy weldable steel is 0·14–0·20%.

It has already been made clear that an unalloyed steel of this carbon range cannot meet the higher tensile requirements and, therefore, in order to develop superior strength suitable alloy additions must be made. Again, limiting factors such as cost of manufacture and production difficulties must be taken into consideration. However, there are a number of possible alloying elements available, e.g., manganese, silicon, nickel, chromium and molybdenum. Various combinations of these alloying elements may now be discussed.

Manganese.—Being the cheapest of the above alloys, manganese is an obvious choice and has been widely used in the manufacture of steels to B.S.548 (Table I). The carbon content of this steel is usually in the region of 0·25%. For fusion welding, quality B.S.968 specifies a maximum carbon of 0·23%, limits being set to the alloying elements as follows :—

Mn %	Ni % (optional)	Cr % (optional)
1·8*	0·5*	1·0

* The combined percentage of manganese and chromium shall not exceed 2·0%.

In the authors' opinion, the upper limits of the alloying elements are too high for good weldability. The Ministry of Supply Schedule of Carbon and Alloy Steels (S.T.A.5) includes a modification of this steel, designated V.6.B, of which it is stated that it is generally as easily weldable as mild steel. As a rule these are supplied as carbon manganese steels, typical analyses and mechanical properties being shown in Table I. Manganese contents above 1·5% should be avoided as can be seen from Fig. 3 which shows the effect of manganese on the mechanical properties of a 0·22% carbon steel. When the manganese exceeds 1·5%

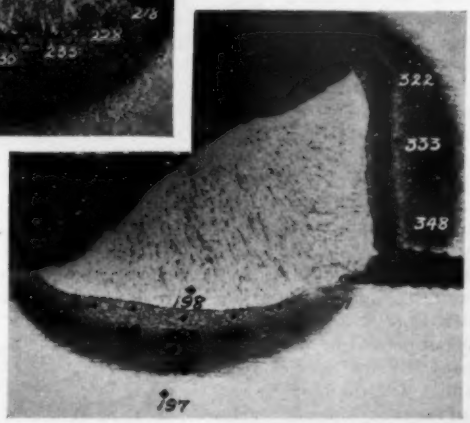


Fig. 6.—C 0.22%, Si 0.72%, Mn 1.40% ½ in. thick plate, 8 SWG. electrodes.

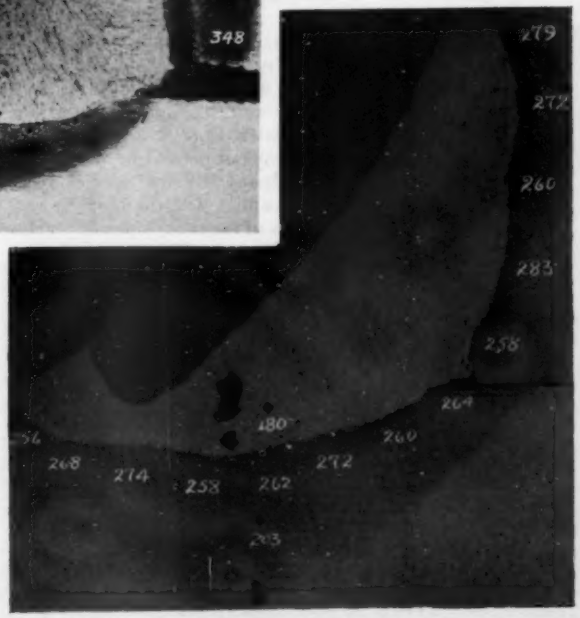


Fig. 7.—C 0.22%, Si 0.72%, Mn 1.40%, ½ in. thick plate, $\frac{5}{32}$ in. dia. electrodes.

serious loss in ductility takes place. Steels of composition in the region of that shown in Table I should give little difficulty in welding provided reasonable precautions are taken.

The controlling factors are plate thickness and electrode gauge. The thicker the plate the more rapidly is heat removed from the welding zone and, consequently, the more drastic is the quenching. The heavier the electrode gauge the greater will be the heat input and, consequently, for a given plate thickness the less drastic will be the quenching. Barr[3] has shown that as the electrode gauge is increased, the maximum hardness developed in a ½ in. thick plate is decreased and that, in order to minimise hardening of the thermally disturbed zone in steels of the type under discussion, it is advisable to use the maximum possible gauge of electrode. Where fillet welds are made on plate thicknesses greater than ½ in. the minimum gauge of electrode should be ¼ in. diam., and as the plate thickness increases, the electrode gauge should be increased up to ⅜ in. diam. Typical cross sections of Reeve tests on carbon manganese steels shown in Figs. 4 and 5 illustrate the beneficial effect of using large gauge electrodes on a ½ in. plate. Attention is also drawn to the memorandum on the "Arc Welding of High Tensile Low Alloy Structural Steels" prepared by the Welding Research Council and published by the Institute of Welding. (Publication T.2).

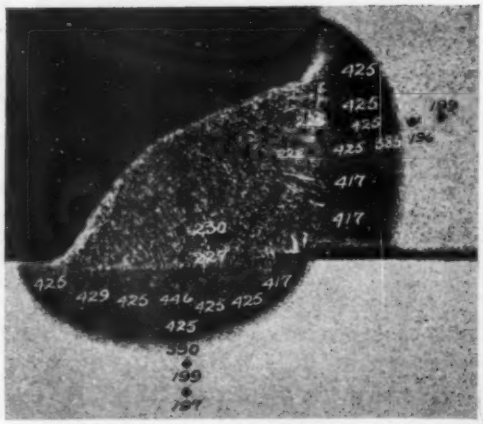
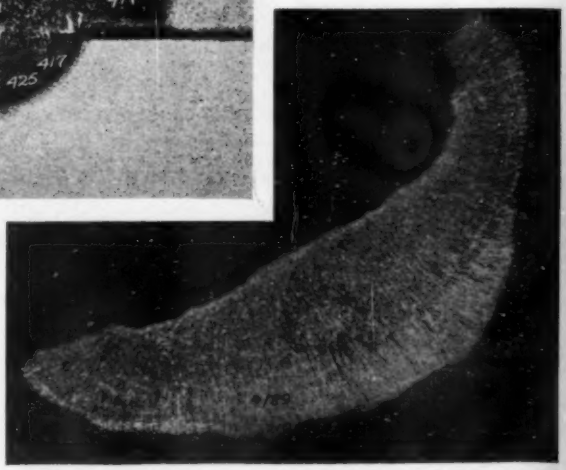


Fig. 8—C 0.20%, Mn 0.78%, Ni 2.4%, ½ in. thick plate, 8 SWG. ← electrodes.

Fig. 9—C 0.20%, Mn 0.78%, Ni 2.4%, ½ in. thick plate, ⅛ in. dia. electrodes. ↓



Silicon.—A review of the literature states that there "is an obvious paucity of mechanical test results on welded silicon steel." With the war, however, and the resulting shortage of other alloying elements, the possibilities of silicon steels with and without manganese has had to be considered. In Germany, silicon found widespread application as an alloying element. It was claimed that silicon gives less weld hardening, but when Figs. 6 and 7 are compared with Figs. 4 and 5 it does not appear that there is any appreciable difference between the maximum hardness set up in the high silicon steel of 0·22% carbon and 1·4% manganese and that in a normal carbon-manganese steel. The mechanical properties of this type of steel are similar to those of a carbon manganese steel*. It does not appear that the high silicon steel justifies the claims made for it, and in addition the high cost and difficulties in manufacture limit the application of such a steel in structural work.

Nickel.—Structural steel plates containing up to 2·5% of nickel find widespread application in industry and have been used in the building of locomotive boilers, the carbon content being limited to 0·2% maximum. Figs. 8 and 9, illustrating cross sections of typical Reeve tests, show that this material when welded even with heavy gauge electrodes gives a degree of hardening of the same order as that of carbon manganese steel of similar mechanical properties. The cost of steels of this type and the precautions necessary to ensure satisfactory welds limit its application to boiler construction and other special purposes.

Chromium.—Reference has been made to unusual alloy combinations used in America. One such steel is Yoloy which depends for its strength on precipitation hardening. This steel contains nickel and copper and has considerable popularity in the United States. Another steel is "Cor-ten" which is now manufactured in this country and finds considerable application where corrosion resistance in addition to a high yield point is required. This steel is of the silicon, chromium, phosphorus, copper, type and has a low carbon content which minimises weld hardening. The high phosphorus

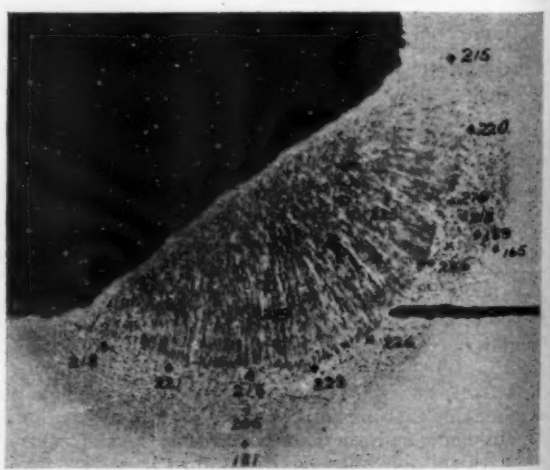


Fig. 10.—C 0.07%, Mn 0.35%, Ni 0.25%, Cr 0.95%, Cu 0.30%, P 0.10%, ½ in. thick plate, 8 SWG. electrodes.

3 *Journal West of Scotland Iron and Steel Institute*, vol. xvi., p. 86.

* The composition and mechanical properties of all the steels discussed are shown in Table II.

does not affect the weldability of the material. In a Reeve test on ½ in. plate the maximum hardness recorded was 226 as illustrated in Fig. 10. Butt welds are satisfactory and if a high-tensile type of electrode is used a deposit can be obtained which has a higher tensile strength than the parent plate. An additional advantage of this steel is the fact that the atmospheric corrosion resistance is much superior to mild steel.

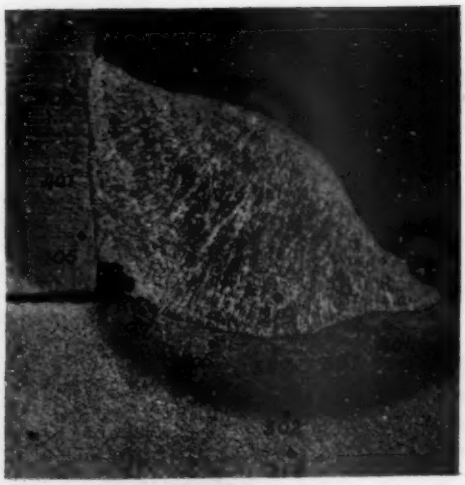


Fig. 11—C 0.18%, Mn 0.54%, Ni 1.95%, Cr 0.22%, Mo 0.24%, ½ in. thick plate, 8 SWG. electrodes.

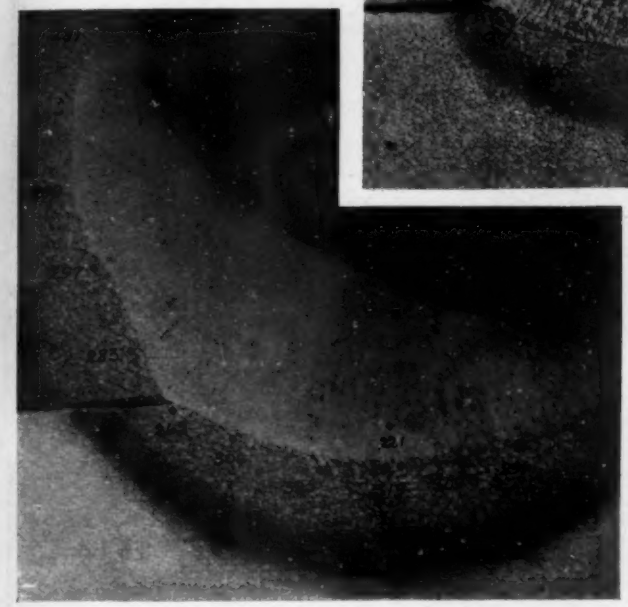


Fig. 12.—C 0.18%, Mn 0.54%, Ni 1.95%, Cr 0.22%, Mo 0.24%, ½ in. thick plate, $\frac{1}{16}$ in. dia. electrodes.

Titanium and Vanadium.—Another line of attack in the development of low alloy structural steels depends on additions of certain elements such as titanium and vanadium to reduce the activity of the carbon by combining preferentially with it, which thus reduces the hardening effect due to carbon. The effect of titanium has been discussed by Swinden and Reeve[4]. Owing to the high cost and the difficulties associated with the manufacture of titanium steels, these have not yet been produced on a commercial scale. Vanadium in conjunction with manganese has, however, been successfully employed.

Molybdenum.—Although it is an expensive element the effect on the mechanical properties as an alloy addition to nickel or manganese steels justifies the consideration of molybdenum. Molybdenum has been used without other alloying elements but the increased molybdenum required to give high yield strength makes this steel expensive and limits its application to hightemperature service where resistance to creep isimportant.

The addition of molybdenum to a 2% nickel steel increases the yield point without a corresponding increase of tensile strength. The results of Reeve tests show that in a ½ in. plate the maximum hardness compares favourably with that of a straight carbon manganese steel

4 *Transactions Institute of Welding*, 1938, vol. 1, p. 10.

Fig. 13.—C 0.18%, Mn 0.54%, Ni 1.95%, Cr 0.22%, Mo 0.24%, ¾ in. thick plate, 8 SWG. electrodes.

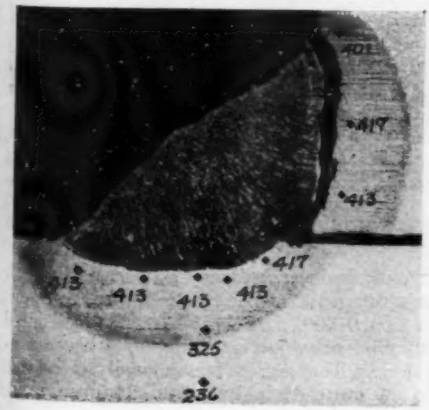
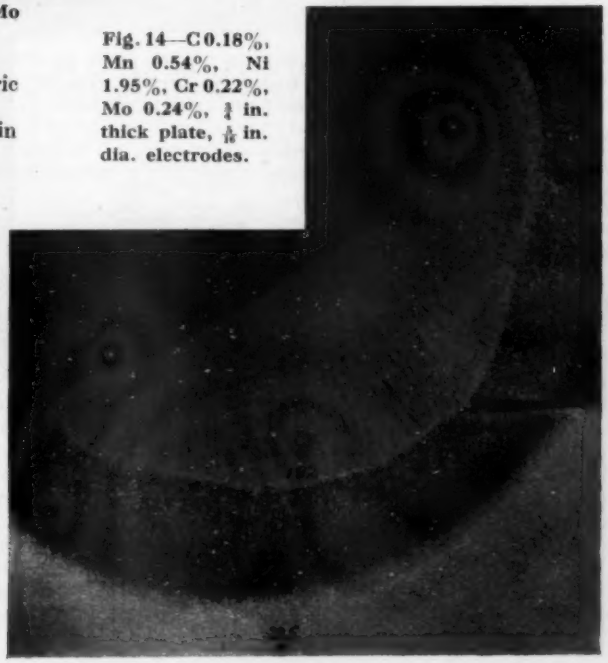


Fig. 14—C 0.18%, Mn 0.54%, Ni 1.95%, Cr 0.22%, Mo 0.24%, ¾ in. thick plate, $\frac{1}{16}$ in. dia. electrodes.



when welded with small gauge electrodes. On using electrodes of larger gauges, the maximum hardness obtained is slightly higher than that of a straight carbon-manganese steel. As the thickness of plate is

TABLE III.—MECHANICAL PROPERTIES OF UX1 STEEL

Plate Thickness	Yield Point Tons/sq. in.	Max. Stress Tons/sq. in.	Elong. % 8 in.	Red. of Area %	Izod Impact ft./lbs.
¼ in. Length	28·6	39·3	16	47·0	—
¼ in. Cross	28·5	38·6	15	46·0	—
1 in. Length	28·2	35·1	18	53·0	34, 41, 38
1 in. Cross	30·0	36·5	20	43·0	23, 24, 27
1½ in. Length	31·7	40·8	19	48·0	42, 38, 48
1½ in. Cross	30·0	41·0	19	42·0	25, 30, 26
3 in. Length	27·7	39·3	19	46·0	34, 40, 28
3 in. Cross	26·4	38·4	18	43·0	29, 38, 28

increased to ¾ in., the hardness set up in the thermally disturbed zone is correspondingly increased. This is shown in Figs. 11 to 14.

Since the cost of manufacture is an important item, the combination of molybdenum with manganese in place of nickel has received considerable attention. S.T.A.5 referred to above includes one such steel (V.8), but the manganese content (1·8 max.) excludes it from the class of easily weldable steels.

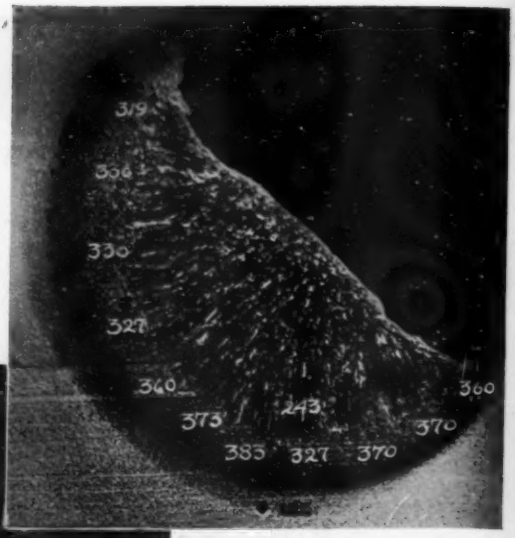


Fig. 15.—C 0.16%, Mn 1.09%, Ni 0.45%, Cr 0.07%, Mo 0.265%, ¾ in. thick plate, 8 SWG. electrodes.

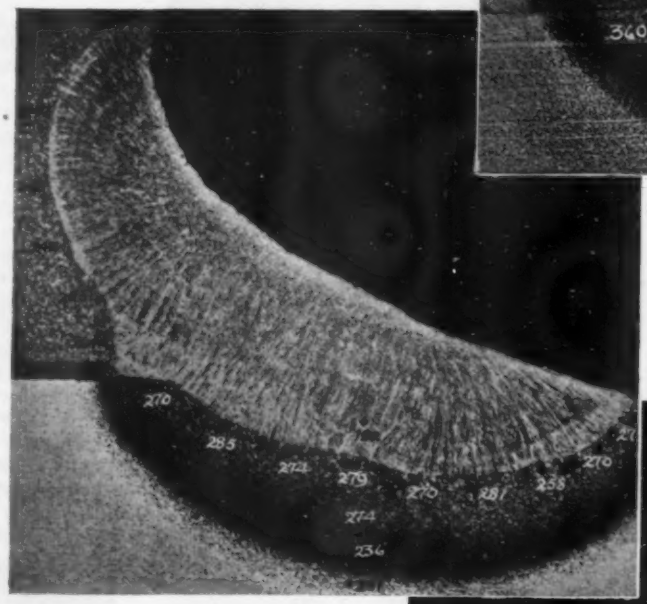


Fig. 16.—C 0.16%, Mn 1.09%, Ni 0.45%, Cr 0.07%, Mo 0.265%, 1 in. thick plate, ⅛ in. dia. electrodes.

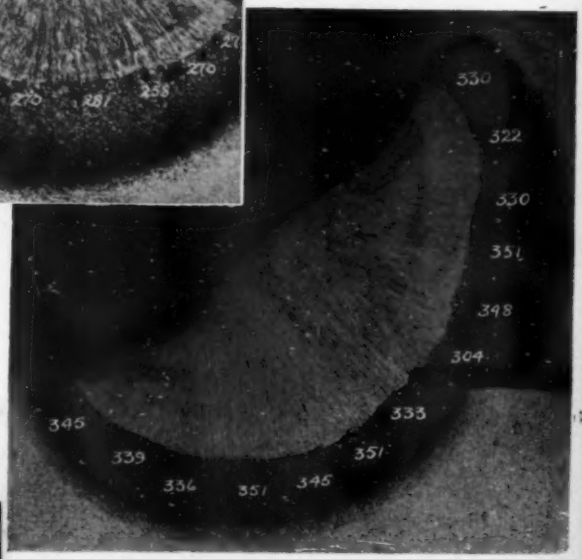


Fig. 17.—C 0.16%, Mn 1.09%, Ni 0.45%, Cr 0.07%, Mo 0.265%, 1½ in. thick plate, ¼ in. dia. electrodes.

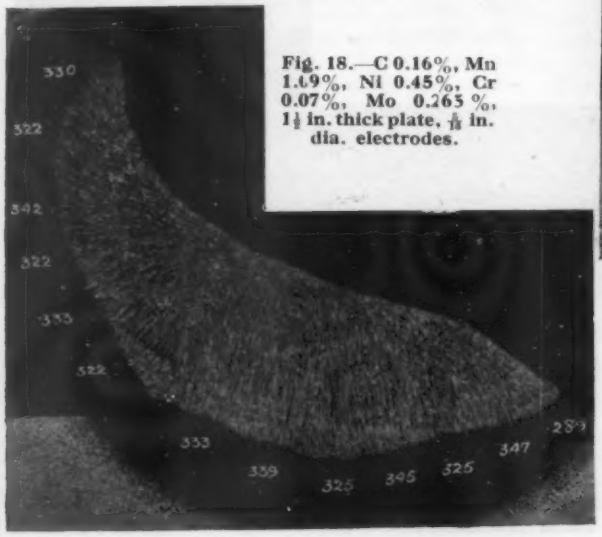


Fig. 18.—C 0.16%, Mn 1.09%, Ni 0.45%, Cr 0.07%, Mo 0.265%, 1½ in. thick plate, ⅛ in. dia. electrodes.

An Easily Weldable Manganese Molybdenum Steel

Most of the steels described above have one or both of the disadvantages that the maximum carbon content is more than 0·20% and that the guaranteed yield point is not sufficiently high for many modern requirements. A demand has arisen for an easily weldable steel with a guaranteed yield point of 25 tons/sq. in.

TABLE IV.—EFFECT OF LOW TEMPERATURE TREATMENT ON A 1 in. THICK PLATE OF UX1 STEEL

Tempering Temperature	Yield Point Tons/sq. in.	Max. Stress Tons/sq. in.	Elong. % 8 in.	Red. of Area %	Izod Impact ft./lbs.
500° C. 1 Hour	28·0	37·0	19	48·0	35, 31, 34
550° C. 1 Hour	28·8	36·8	20	49·0	38, 34, 39
600° C. 1 Hour	29·0	36·2	20	51·0	42, 36, 40
650° C. 1 Hour	25·4	34·2	24	54·0	50, 45, 47

As a result of extensive research carried out in the authors' laboratories under the direction of Mr. William Barr, Chief Metallurgist, a manganese-molybdenum steel (UXI) has been produced which satisfies both of these requirements. A typical composition of this steel is as follows :—

C %	Mn %	Ni %	Cr %	Mo %
0·16	1·09	0·45	0·07	0·265

The results of Reeve tests on ½ in., 1 in. and 1½ in. plates illustrated in Figs. 15 to 18, show that the welding characteristics are excellent, the maximum DPH on a 1½ in. plate using a ¼ in. diam. electrode being 351. Typical mechanical properties in the "as rolled" and tempered conditions given in Table III show that the increased yield strength has been obtained without loss of ductility.

Since these low alloy molybdenum-containing steels are mildly air hardening, when the finishing temperature of rolling is controlled they show a favourable response to low temperature treatment. This is shown in Table IV. It will be noted that with reheating temperatures up to 600° C. there is no loss in yield strength and there is a gain in ductility and toughness. Most of the low alloy structural steels suffer from the disability that the yield point obtained varies considerably with the thickness of material. On the other hand, the response to low temperature treatment, together with controlled rolling of UX1. steel enables a minimum yield point of 25 tons/sq. in. to be obtained on all thicknesses up to 3 in. Another advantage is that the effect of thermal softening in the welded zone is negligible. A more complete account of the manufacture and application of this high yield point weldable low alloy steel will be published later.

The foregoing review does not claim to be exhaustive, nor has the last word been said on the subject. Doubtless, like Oliver Twist, the engineer will ask for more and the next step will be the demand for an easily weldable steel with a yield point of 30 tons/sq. in. To paraphrase C. H. Grattan[5], "If all the people who demand high strength with easy weldability were laid end to end it would be a good thing."

5 *Harper's Magazine*, June, 1945, p. 577.

The Sigma Phase in Certain High-Chromium Steels

By Francis B. Foley

Superintendent of Research, The Midvale Company, Philadelphia.

The hard, brittle and non-magnetic constituent formed in certain high-chromium steels is discussed. This constituent has been termed the "sigma phase," and its formation in appreciable amounts will cause an alloy steel to lose its ductility to such an extent that the usefulness of the steel may be seriously impaired. Because of the increasing use of heat-resistant iron-chromium-nickel alloys in modern industrial processes, this discussion on the nature and occurrence of the sigma phase, by the chairman of the Technical Research Committee of the Alloy Casting Institute of America, is opportune. He presents an up-to-date history, and at the same time a compilation of important results obtained in studies of the occurrence of this brittle constituent.

At first identified as the simple compound FeCr, it has been found that the composition of the sigma phase may be extended to take in other elements, notably nickel, molybdenum, silicon, vanadium, manganese and aluminium, without affecting its crystallographic form. The phase is stable below about 925° C., but at higher temperatures it transforms to magnetic alpha (ferrite). The transformation is a sluggish one. Sigma may be the cause of serious embrittlement, particularly after long exposures in a suitable range of temperatures.

In 1927 Bain and Griffiths[1] then of the Union Carbide and Carbon Research Laboratories, presented a paper, "An Introduction to the Iron-Chromium-Nickel Alloys," before the A.I.M.E., in which they described a phase containing from 43 to 53·6% Cr, 0·3 to 10·3% Ni, and C 0·15 to 0·44%, which they called the "B" constituent. This "B" constituent was described as :

(*a*) Massive form—a clear-etching, corundum-hard, brittle, non-magnetic constituent, usually filled with faint cracks ;

(*b*) Troostite form—dark-etching solid dispersion of sub-microscopic particles of true "B" constituent in various amounts of ferrite matrix. Weakly magnetic, hard and often cracked. An intermediate form in ferrite changing to "B" constituent by long-time heating at low temperature.

In discussing the effect of the brittle "B" constituent, they record the investigation by X-ray of a tube embrittled after several months at 550° C. (1,022° F.), composition not given, which brought out "in addition to a strong ferrite pattern," 12 clear supernumerary lines checking perfectly the first 12 strong lines of the "B" constituent. In this case the brittle constituent was shown microscopically as a grain boundary constituent. The limiting compositions for the "B" constituent were not definitely fixed by Bain and

1 Bain and Griffiths, Trans. A.I.M.E., 1927, vol. 75, p. 166.

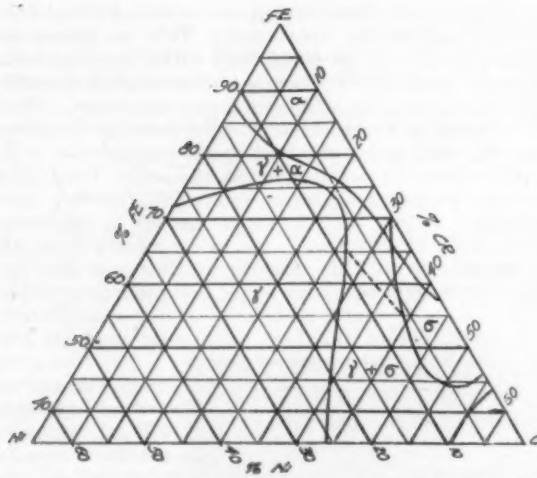


Fig. 1.—Iron-chromium-nickel system showing phase locations after slow cooling from 875° C. (Bain and Griffiths, Trans. A.I.M.E., 1927, vol. 75, p 166.)

Griffiths. They detected its presence in alloys containing as little as 30% of chromium, and observed that the alloy is already very brittle before the extent of the constituent is observable microscopically.

Bain and Griffiths finally concluded that this new phase (" B " constituent) develops at some temperature above 900° C., and probably below 950° C. There is no doubt that in this work they described, for the first time, the sigma phase which has been subsequently the subject of study by a number of investigators. Their diagram of the iron-chromium-nickel system, showing phase locations after slow cooling from 875° C., is reproduced in Fig. 1.

In 1931 Wever and Jellinghaus[2] showed this brittle constituent of Bain and Griffiths to be the compound FeCr, which they found capable of dissolving surplus iron and chromium. However, they failed to take into account the extreme sluggishness with which this phase forms and transforms. Due to the fact that they did not use long-time exposures, they failed to produce true equilibrium. Their phased diagram is shown in Fig. 2. S. Eriksson[3] in 1934 published in *Jernkontorets Annaler* confirmation of the existence of the intermediate phase corresponding to FeCr in a study of the binary Fe-Cr and Cr-Ni systems.

Jenkins[4] and his co-workers made no reference to the sigma phase in their work on the ternary Fe-Cr-Ni system reported in the *Journal of the Iron and Steel Institute*, 1937. This was not because they did not allow sufficient time for it to form, having annealed specimens for over 31 days at 800° C., but probably because they made neither hardness, magnetic nor X-ray determinations. The embrittling effect, as Bain and Griffiths observed, often precedes microscopical evidence of the presence of sigma, so that metallographic detection is not always possible. Six of the 31 alloys studied by Jenkins *et al* were capable of developing sigma. One alloy in particular, having 48·3% Cr, 8·9% Ni, should have contained large quantities of sigma after exposure of 21 days at 800° C. According to Schafmeister and Ergang, this alloy should have been converted entirely to sigma after this treatment. The nearest composition studied by Schafmeister and Ergang had 45·4% Cr, 8% Ni, and was pure sigma after 200 hrs. at 800° C. Jenkins remarked that none of their alloys " lies at the composition of the compound mentioned above " (FeCr), and they made no attempt to investigate changes occurring below 800° C. Presumably, they considered they were free of sigma complications, and did not look for this phase.

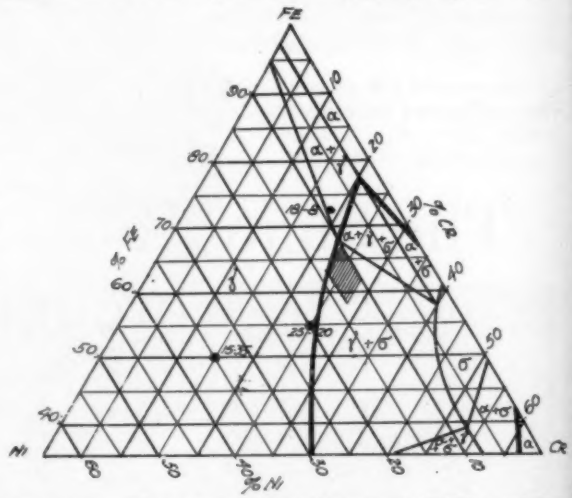


Fig. 3.—Iron-chromium-nickel diagram at 650° C. (Schafmeister and Ergang, Archiv für das Eisenhuttenwessen 12, No. 9, Mar. 39, p. 459.)

The work of Schafmeister and Ergang[5] on the ternary system Fe-Cr-Ni, was published 1939. The carbon content of their alloys was between 0·02 and 0·14, usually 0·05 with only one alloy having C over 0·10%. Conditions of equilibrium as they found them at 800° C.

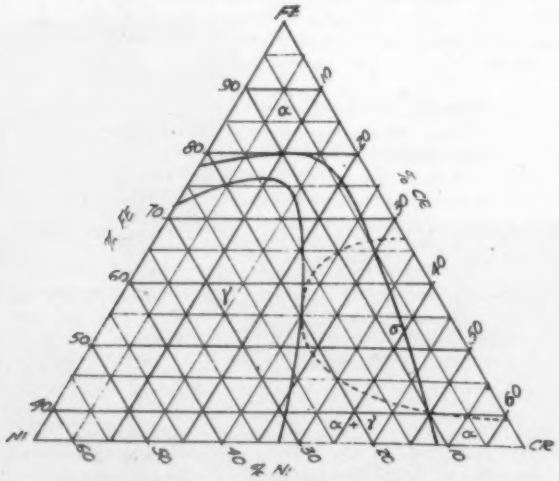


Fig. 2.—Iron-chromium-nickel system. (Wever and Jellinghaus, K.W. Inst. für Eisenforschung, 1931, vol. 13, pp. 93–108.)

2 Wever and Jellinghaus, K.W. Inst. für Eisenforschung, 1931, vol. 13, pp. 93 and 143.

3 S. Eriksson, Jernkontorets Annaler, 1934, vol. 118, p. 530.

4 Jenkins, Bucknall, Austin and Mellor, *Jrn. Iron and Steel Institute*, 1937, No. 11, p. 187p.

5 Schafmeister and Ergang, Archiv für das Eisenhuttenwessen 12, No. 9, Mar., 1939, p. 459.

are shown in Fig. 3. Sigma phase was found in alloys having as much as 33% Ni (with 37% Cr), and in alloys having Cr as low as 19% (with 2% Ni), and as high as 65% (with 10% Ni). The boundaries of the area within which sigma is a stable phase are drawn in heavy lines in the diagram. Specimens were annealed at 650° C. and 800° C. for 200 hrs. and 1,000 hrs. Alloys were examined microscopically, magnetically and by X-ray and hardness testing. The hard, non-magnetic characteristics of the brittle (sigma) constituent were verified. The single-phase sigma area, extending inward in Fig. 3 from the Fe-Cr side at 39–51% Cr, crystallises from the molten state as alpha and transforms, according to Schafmeister and Ergang, to sigma upon sufficiently long exposure at temperatures below about 920° C. (1,688° F.), as stated by Bain and Griffiths. Nickel is soluble in this phase in amounts up to about 10%. The region of sigma stability, bounded by heavy lines (Fig. 3) can be shifted by impurities. These investigators found that, whereas an alloy of C 0·07, Si 0·32, Cr 24·8, Ni 19·5 showed no precipitation of sigma after 1,000 hrs. at 800° C., a similar alloy with 3% Si had a strong showing of sigma after the same treatment. Andersen and Jette[10] showed this same effect of Si in the Fe-Cr-Si system, in which the field of sigma is extended to lower chromium contents with increasing silicon. Schafmeister and Ergang make the interesting observation that with about 25% Cr, 20% Ni, 2·5% Si there was considerable precipitation of sigma and low notch sensitivity (0·7–1·2 Mkg/cm²) after 1,000 hrs. at 800° C. when carbon was increased from 0·06–0·32%. When Si was reduced to 0·77% (0·11% C), notched resistance increased to 3·3 Mkg/cm².

Bradley and Goldschmidt[6], working on the same system, published results in 1941 which agree generally with those of the two German investigators, but with phase boundaries widely different in many places. This shows the susceptibility of the alloys in this system to heat-treatment. Bradley and Goldschmidt depended largely on X-ray diffraction work, a method which they believe sufficient in itself to establish phase boundaries. They found the single-phase sigma region of the diagram, Fig. 4, to extend to a maximum solubility of 6% nickel, and to cover from about 45% to nearly 60% atomic Cr, and remark that it is very difficult, due to sluggishness of reactions in the Fe-Cr-Ni system, to attain equilibrium. For accurate delineation very long periods of heat-treatment are required. They found the lines of the sigma phase X-ray pattern not always identical, and suggested the possibility of two modifications of sigma, one stable, the other a metastable, possibly transitional phase, depending for its structure on whether the original solid solution was face-centred or body-centred cubic. This suggestion is somewhat at variance with the commonly-accepted view, expressed by Schafmeister and Ergang, that the sigma constituent forms entirely from ferrite. Bradley and Goldschmidt found the majority of the lines for the sigma phase distributed around positions corresponding to a body-centred cubic lattice, reminiscent of the lines produced by alpha-manganese.

An example of the sluggishness of the alpha-sigma transformation appears in the work of Jette and Foote[7], who found that powdered 50·6 atomic per cent. Fe-Cr alloy, which showed only a trace of sigma after 250 hrs.

6 Bradley and Goldschmidt, *Jrn. Iron and Steel Institute*, 1941, No. 11, p. 273p.
7 Jette and Foote, *Metals and Alloys*, 1936, vol. 7, p 207.

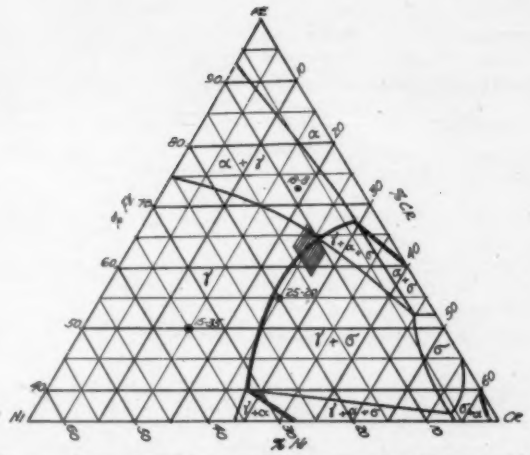


Fig. 4.—Iron-chromium-nickel system based on specimens cooled at 10°–20° C. per hour from 900° C. after a variety of previous heat-treatments. (Bradley and Goldschmidt, "Jrn. Iron and Steel Institute," 1941, No. 11, p 273p.)

at 600° C. was completely converted to sigma in 24 hrs. at the same temperature after it had been preliminarily cold-worked by light rubbing in an agate mortar.

In 1943 Cook and Jones[8], of the Alloy Steels Research Committee of the Iron and Steel Institute, published the results of their work on the binary Fe-Cr system using high-purity alloys of 25–76% chromium. Fig. 5 shows the extent of the sigma phase as they found it by the powder method of X-ray technique in conjunction with microscopic examination of solid samples. These authors do not mention the compound FeCr anywhere in their paper. In the range 750°–830° C. they experienced considerable difficulty, because of the extremely sluggish character of the transformation in obtaining equilibrium in alloys of 43 up to 50 atomic per cent. of chromium. An alloy of 46·8 atomic per cent. of chromium was prepared in the alpha, in the sigma, and in the alpha + sigma states. Annealing for 17 days at 800° C. produced no detectable change. Similar specimens resisted transformation in seven days at 780° C., and in 26 days at 810° C. Two specimens of the same alloy, one pure sigma, the other alpha +

8 Cook and Jones, *Jrn. Iron and Steel Institute*, 1943, No. 11, p. 217p.

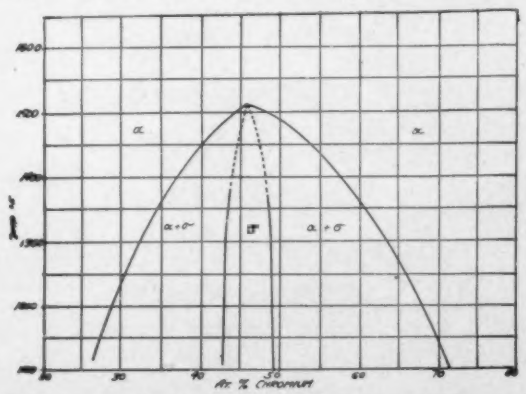


Fig. 5.—Iron-chromium phase diagram over the range 25–76% chromium. (Cook and Jones, "Jrn. Iron and Steel Institute," 1943, No. 11, p. 219p.)

sigma, transformed to the two-phase structure in comparable phase proportions after only seven days at 820° C. Increasing the time at 820° C. to 24 days caused a transformation of both the alpha + sigma, and the pure sigma to pure alpha. From this it is concluded that the upper limit of stability of the sigma phase in the Fe-Cr system is 820° C. This is considerably below the temperature of about 920° C. suggested by Schafmeister and Ergang for the Fe-Ni-Cr system. At 830° C. Cook and Jones found that pure sigma transformed entirely to alpha in 11 days, while an alloy of 48·8% Cr, which had been sigma, gave a 95% alpha structure after only 1 hr. at 840° C. There is evidence that prior cold-working accelerates the attainment of equilibrium upon subsequent exposure to elevated temperatures.

Characteristics of the Sigma Phase

Summarising the findings from the most important papers dealing with sigma phase in the Fe-Cr and the Fe-Cr-Ni systems, certain characteristics of the sigma phase seem to have been reliably established.

1. Sigma appears to be the intermetallic compound FeCr in which the chromium, in Fe-Cr alloys richer in iron than 50 atomic per cent., may be replaced by iron atoms. In the binary Fe-Cr system, this extends the range of pure sigma to cover chromium contents between 44 and 50 atomic per cent.

2. In the binary Fe-Cr system, sigma persists as a phase in equilibrium with alpha in the chromium-rich alloys containing up to 71 atomic per cent., and in the Fe-rich alloys with chromium contents as low as 26 atomic per cent.

3. Sigma recrystallises in the simple body-centred cubic lattice structure of alpha in Fe-Cr alloys at temperatures above about 820° C. The reverse transformation from alpha to sigma is very sluggish, making it difficult to fix the boundaries definitely with respect to both temperature and alloy concentration.

4. Sigma is a hard, brittle, non-magnetic phase. It has a complicated body-centred cubic lattice structure.

5. With the addition of nickel to the Fe-Cr system the sigma phase persists at temperatures, up to 920° C., and its limits are extended. It is then found in equilibrium with non-magnetic gamma as well as with magnetic alpha. According to Bradley and Goldschmidt, sigma may be found in alloys containing as much as 32% Ni, with 28% Cr and 40% Fe (all atomic percentages). With the minimum of nickel at 3%, the range of chromium content of alloys in which sigma may be present extends from 29% to nearly 95%. When nickel is introduced into the system it appears to replace iron in the FeCr compound so that the composition of pure sigma decreases in Fe content from 60%, as a maximum without nickel, to 39% at its leanest, while nickel is increasing from zero to 8%, according to Schafmeister and Ergang. The same occurrence appears in the Bradley and Goldschmidt diagram wherein the iron content of pure sigma is decreased from a range of 45 to 52 atomic per cent. maximum with no nickel to 36 per cent., as nickel increases from zero to 6%.

Cook and Jones found sigma to be a stable phase in the range 26–71% Cr in the binary system Fe-Cr. The ternary Fe-Cr-Ni diagrams (Figs. 3 and 4) should probably be re-examined in the light of this latest work which, by altering the Fe-Cr side of the diagrams, would affect considerably the areas within them.

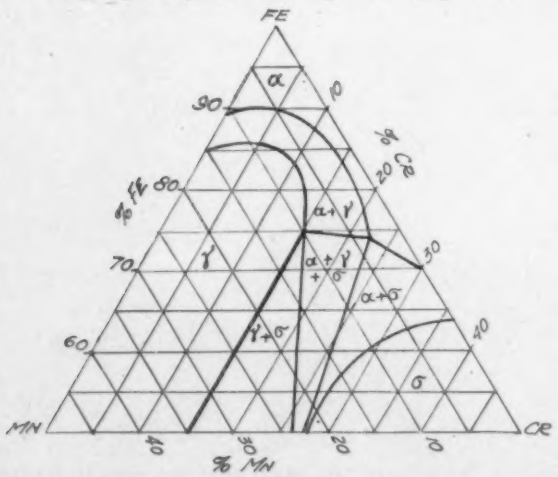


Fig. 6.—Iron-chromium-manganese diagram of 650°C. (Burgess and Forgeng. Trans. A.I.M.E., vol. 131, 1938, p. 277.)

Presence of Sigma in Other Systems

While the occurrence of sigma in Fe-Cr and Fe-Cr-Ni alloys is commercially of most importance, its occurrence in other ferrous alloy systems is also of interest. In 1938 Burgess and Forgeng[9] studied alloys of Fe-Cr-Mn with C below 0·16% and Mn and Cr up to 50% (Fig. 6). The behaviour of Mn was like that of Ni in the Fe-Cr-Ni system. With Mn of 17% and higher, sigma was found in alloys containing as little as 15% chromium. In the Fe-Cr-Ni system, Schafmeister and Ergang found the sigma phase to extend only to as low as about 19% Cr with 2% Ni. Burgess and Forgeng found Mn to accelerate remarkably the ferrite to sigma transformation. In an alloy of 44·91% Cr with 0·79% Mn, it required seven hours at 700° C. to transform alpha to sigma, whereas another alloy with nearly the same chromium (44·82%), but having 5·34% Mn transformed in one hr. They found the crystal pattern of the sigma phase in 30% Fe-Cr alloy containing only 0·71% Mn. They however, consider it a mistake to designate this phase as a compound in view of the manner in which Mn, N Si, etc., may be substituted in it. The hardness of the sigma phase in the Fe-Cr-Mn system was given in terms of scratch micro-hardness as 3906–5917 K, while that of Cr_4C was 2000–3906 K, and of ferrite 657–977 K.

Andersen and Jette[10] described in 1936 the occurrence of sigma in the Fe-Cr-Si system. With Si at 14·5%. sigma was found in the alloy containing as little as 15·6% Cr, and with Si 28·6% sigma appeared in the 7·26% Cr alloy. They also describe the phase as hard and brittle and a secondary solid solution of three components based on the Fe-Cr phase of the iron-chromium binary system. They observed that Si seemed to accelerate the precipitation of sigma from the alpha phase. An alloy with only 0·90% Si was distinctly slower in reaching equilibrium than alloys containing Si 2·32, 2·88 and 2·85. In a later paper[11] (1938) these authors gave consideration to the form of

9 Burgess and Forgeng, Trans. A.I.M.E. (I. and S. Div.), vol. 131, 1938, p. 277.
10 Andersen and Jette, Trans. A.S.M., vol. 24, 1936, p. 375.
11 Jette and Andersen. Trans. A.I.M.E. (I. and S. Div.), vol. 131, 1938, p. 303.

Fig. 7.—Sigma (containing cracks) in 12% Ni, 27% Cr 1.4% Mo after 1,273 hrs. at 982° C. under 1,000 lbs. sq. in. load. × 500. (Photomicrograph by H. S. Avery, Amer. Brake Shoe Co.)

the carbides in the ternary Fe-Cr-Si system. A discontinuity in the plot of per cent. Cr vs. the alpha lattice parameter was found at C 0·35 and 0·50, depending on the gross alloy content. The carbide is based on the compound Cr_7C_3 for carbon contents higher than the point of discontinuity. Therefore, it was observed that the composition of the alpha phase matrix is easily changed by addition of C, and consequent removal of Cr from solution in the matrix by the formation of Cr_7C_3, to the point where it will remain as alpha at all temperatures. With carbon exceeding 0·3–0·4%, the sigma phase could not be found by any means. In another paper[13] dealing largely with micro-structure of the Fe-Cr-Si alloys, these investigators showed sigma to appear microscopically in a variety of forms depending on the conditions under which it crystallised, and that the mode influenced considerably the hardness. Sometimes it occurred, along with alpha, in patterns suggesting a eutectic, sometimes it was coarsely lamellar and partly globular, sometimes irregularly distributed in the form of sharp-cornered elongated areas. Quenched and aged sigma-alpha structure sometimes developed a distinctly rectangular Widmanstatten (martensite) structure. Solution quenched from 1,025° C. or 1,100° C. with 18-hrs. soaking time, and aged for 12–52 hrs. at 700° C. resulted in increasing hardness of as low as 222 Brinell up to from 340–509 Brinell. Hardening was believed to be due to precipitation of sigma in finely-divided form. Andersen and Jette state that there can be no doubt that in both ternary systems (Fe-Cr-Si and Fe-Cr-Ni) "we have to do with an iron-chromium compound with a composition in the neighbourhood of FeCr, which is capable of dissolving not only excess iron and chromium, but also silicon and nickel." Wever and Jellinghaus[14] found this same structure in the Fe-V system.

Occurrence of Sigma in Commercial Alloys

The positions of four typical commercial alloys—*i.e.*, 18 Cr-8 Ni, 25 Cr - 12 Ni, 25 Cr - 20 Ni, and 15 Cr - 35 Ni, are shown in the interesting portion of the ternary diagrams according to Schafmeister and Ergang (Fig. 3) and Bradley and Goldschmidt (Fig. 4). The 15 Cr-35 Ni alloy falls comfortably within the single-phase gamma field. Note that the three alloys in which chromium predominates over nickel either fall within or very close to areas in which sigma is a stable phase. In both diagrams 18 Cr-8 Ni is in a two-phase alpha-gamma area, but not so far away, particularly in the Schafmeister and Ergang diagram, from an area in which sigma is stable that a shift in composition together with the addition of some sigma-promoting element, such as Mo, Si or Al, could not easily cause this phase to form. Both diagrams agree in placing 25 Cr-20 Ni close to the border between the single-gamma phase area and the two-phase gamma-sigma region. As already pointed out, Schafmeister and Ergang found that, although no sigma appeared in a 25 Cr-20 Ni alloy containing Si 0·32, strong sigma lines developed when Si was increased to 3%.

The area covered by the A.S.T.M. specifications for 25 Cr-12 Ni alloy is contained within the rectangulararea in the two diagrams. At 650° C. Schafmeister and Ergang's diagram has this area wholly within regions in which sigma is a stable phase. According to Bradley and Goldschmidt most of the area lies within the two-phase sigma-gamma region. Both these diagrams are based on as nearly pure materials as the investigators commanded. Carbon is thus as low as possible under the circumstances, usually well below 0·10%.

Recent work of the American Alloy Casting Institute has added further information concerning the incidence of sigma in the commercial 25–12 alloy. The importance of this work is evident from the position of the 25–12 alloy in the ternary Fe-Cr-Ni system. Bearing in mind that carbon is very low in both cases, the effect of changing the concentration of nickel and of chromium of the nominally 25 Cr-12 Ni alloy on its position in the ternary system can be observed in both the Schafmeister and Ergang and the Bradley and Goldschmidt diagrams. In the former, which indicates equilibrium at 650° C.,

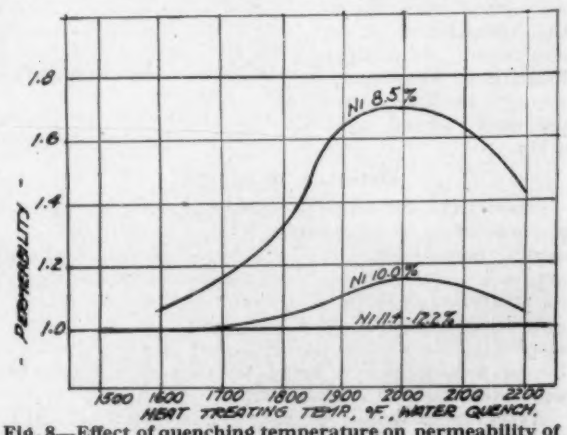


Fig. 8—Effect of quenching temperature on permeability of 25 Cr-12 Ni alloy and of varying nickel content. Base composition C 0.31–0.39, Mn and Si 0.40–0.60, Cr 25.7–26.4. (After Avery, Cook and Fellows. Trans. A.I.M.E., vol. 150, p. 392.)

12 Gow and Harder. Trans. A.S.M., vol. 30 (1942), p. 855.
13 Andersen and Jette, Trans. A.I.M.E. (I. and S. Div.), vol. 131, 1938, p. 218.
14 F. Wever and W. Jellinghaus, K.W. Inst. Eisenforschung (1930) 12, 317.

the shaded area, which lies practically wholly within the $\gamma + \sigma$ field, would move into the single-phase γ field where the nickel kept constant and the chromium decreased by little more than 5%. With chromium constant it would require an increase of the order of 10% in nickel to accomplish the same result. The shaded area in the Bradley and Goldschmidt diagram, which may also be taken to show equilibrium at 650° C. extends into the two-phase regions $\alpha + \gamma$, and $\gamma + \sigma$, and also into the single γ-phase area. With nickel kept constant, a decrease of about 3% in the chromium content takes the alloy out of the sigma region, and into either a wholly γ or an $\alpha + \gamma$ region, depending on the nickel content.

In commercial alloys of the 25–12 type with their usual carbon, manganese and silicon contents, Gow and Harder[12] found that the ratio of nickel to chromium necessary to insure a stable austenitic alloy is expressed by the formula $\frac{\text{Cr\%}-16\ \text{C\%}}{\text{Ni\%}} =$ less than 1·7. Their work produced strong evidence of the effects of Si and Mo on the presence of sigma. Si in excess of 1% had the effect on structure of three times its weight percentage in Cr and Mo, the effect of four times its weight in Cr in the 25 Cr-12 Ni commercial cast alloy. In alloys containing carbon, both Cb and Ti, by forming stable carbides, and thus leaving little or no carbon available for the formation of chromium carbide, aid in sigma formation. Aluminum, a strong ferritiser, also promotes sigma formation.

Fig. 9.—Ductile single-phase Fe-Cr-Mo alloy. × 500.

Fig. 10.—Brittle two-phase Fe-Cr-Mo alloy (Sigma net-work). × 500.

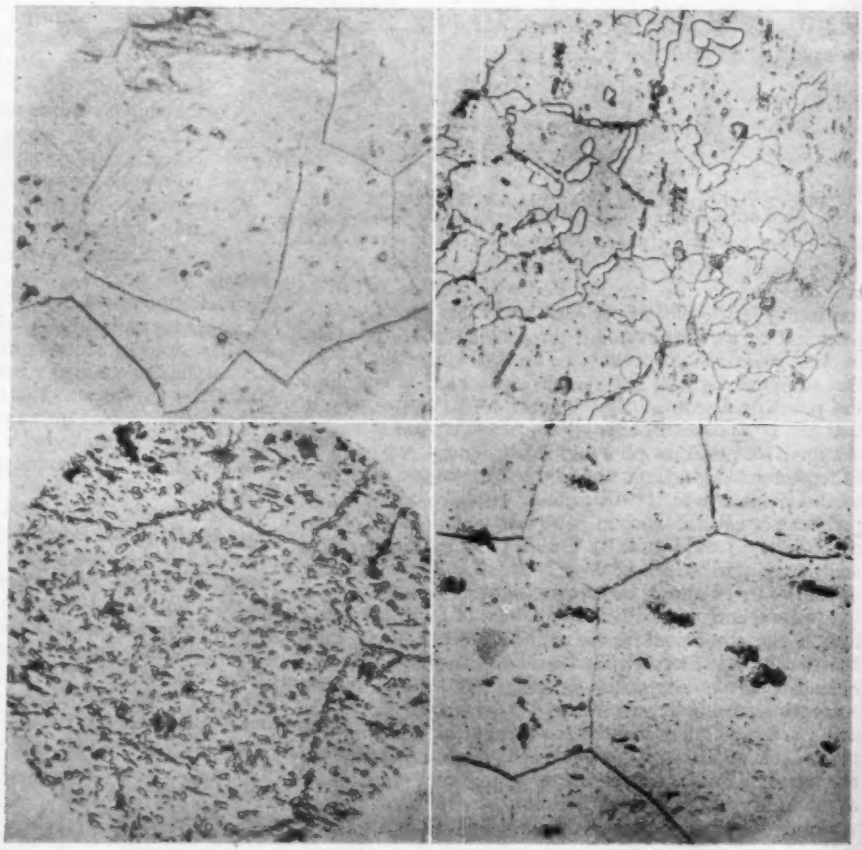


Fig. 11.—Ductile Fe-Cr-Mo alloy air-cooled after 15 min. at 1,600° F. × 500.

Fig. 12.—Brittle Fe-Cr-Mo alloy restored by air cool from 2,000° F. × 500.

Detection of Sigma

Apart from the obvious method of determining the presence of sigma by means of X-ray crystal diffraction, careful magnetic analysis appears to be an excellent approach to the problem of detecting this phase where its presence is otherwise in doubt. When present in considerable quantities, it is readily detected microscopically. In such cases sigma is frequently, though by no means always, found to abound with cracks which terminate at its boundaries (see Fig. 7).

Avery used double etching—equal parts of HCl and water followed by a hot 10% solution of alkaline potassium ferricyanide. To reveal small amounts of sigma, Burgess and Forgeng used a modified Murakami reagent prepared by mixing 30 grms. each of potassium hydroxide and potassium ferricyanide with 60 c.c. of water. This solution is made up fresh, and used at boiling temperature.

The embrittling effect of this phase may manifest itself, as stated before, without its presence being detectable microscopically. In such cases a sensitive permeameter, or X-ray diffraction, may be resorted to. The fact that sigma is a transformation product of alpha and, unlike alpha, is non-magnetic, makes the permeameter a useful tool for detection. In the case of Fe-Cr alloys permeability is decreased by the presence of sigma, and in the case of austenitic Cr-Fe-Ni alloys permeability is increased by the presence of alpha. An austenitic Fe-Ni-Cr type which therefore will be $\gamma + \sigma$ and wholly non-magnetic after long exposure to temperatures of say 760°–870° C., will become $\gamma + \alpha$, and therefore magnetic upon sufficient exposure to temperatures upwards of 930° C. The maximum effect on permeability appears to be produced at 1,100° C. in the 25 Cr-12 Ni type alloy according to Fig. 8, after Avery, Cook and Fellows[15]. In order to prevent reversion of the alpha phase, formed at the high temperature, to

15 Avery, Cook & Fellows. Trans. A.I.M.E. (I. and S. Div.), vol. 150, p. 373.

sigma, during cooling, quenching in water is indicated. The fact that normally magnetic ferritic alloys, such as Fe-Cr and Fe-Cr-Si alloys, have their permeability decreased by the presence of non-magnetic sigma, suggests permeability tests after long soaking in the 760°–870° C. range as a means of detecting its presence.*

The effect of sigma on the physical properties of various types of high-alloy materials is not dealt with in this paper. It may, however, be stated as a generality that sigma is harmful to the extent that low or even complete absence of ductility may not be tolerated under the conditions of service for which a casting or wrought product is destined. The presence of sigma at the temperature of hot-working may be the determining factor in selecting the best temperature for the successful production of wrought products. The transformation to sigma during the slow cooling inevitably involved in the production of castings of large sections may result in cracking if the sigma transformation is extensive. The effects of sigma on the mechanical properties of the 25 Cr-12 Ni type cast alloy are pretty thoroughly covered by Avery, *et al* [15] and by Gow and Harder.[12]

In alpha-sigma alloy of C 0·12, Mn 0·60, Si 0·96, Cr 22·8, Mo 3·35, Ti 1·20, which is highly resistant to a number of chemicals, notably phosphoric acid, shows the kind of ductility which it is possible to develop in spite of the fact that sigma is one of its stable phases. This alloy is produced in the form of ¼ in.-thick rolled plate, and also in the form of cold-drawn wire which is woven into wire screening. The successful avoidance of the embrittling effect of sigma depends on the avoidance of the formation of that phase during fabrication by hot-working. The microstructure of the ductile metal which lends itself readily to cold drawing (Fig. 9) is that of a single-phase alloy—in this case, alpha. Wire bar, too brittle for drawing, has the two-phase structure (Fig. 10) alpha with sigma in grain boundaries. When, in the form of $\frac{1}{16}$-in. wire, the ductile alloy is heated to 870° C. for as short a time as 15 min., and then air-cooled, it becomes brittle with the formation (Fig. 11) of sigma both in grain boundaries and within the crystals. Further heating to 1,800° F. followed by air cooling restores the single-alpha phase structure and ductility. The brittle material is transformed (Fig. 12) to the ductile alpha state after a 1,100° C. air quench. The brittle state of this alloy is produced if, in hot working, the temperature is permitted to fall below 930° C. In that event sigma forms, and is drawn out, in hot-working of plate, into brittle films which may cause cracking when deformed with the production of a laminated fracture. When worked above the temperature at which sigma transforms to alpha, 36 in. × 23 in. × 3½ in. billets are rolled without difficulty to ¼-in. plate with Brinell hardness 187/207. When rolled too-cold, cracking often results, and the hardness is increased to near 300 Brinell.

* This statement applies only to those normally ferritic alloys which would not transform to austenite in the 760°–870° C. range, since the permeability would decline if alpha transformed either to the gamma or to the sigma phase.

Verification of Tensile Testing Machines

By W. Thompson, A.F.R.Ae.S.

One of several methods may be adopted to verify the accuracy of tensile testing machines. Each method has its advantages and disadvantages, but in this article attention is directed to the comparative ease of checking by means of the steel proving-ring, the method for which is described.

IN order to ensure the continued accuracy of tensile test machines, it is essential that they be calibrated at regular intervals. Normally the manufacturer arranges for the first calibration when a new machine is installed but thereafter the responsibility for maintaining the accuracy devolves upon the owner.

Most authorities specify that the accuracy shall be within ± 0·5% of the indicated load throughout the range, and that the machine shall be sensitive to a load not greater than the smallest load which can be recorded by the load measuring device.

Primary Standards

The calibration of relatively low capacity tensile machines is frequently effected by the direct application of standard weights to the weighing mechanism, and providing certain simple precautions are taken this is the most accurate method. It is, of course, necessary to arrange the weights so that the centre of gravity of the load lies in a vertical line through the centre of the plattens, and if a yoke is used it must be symmetrical.

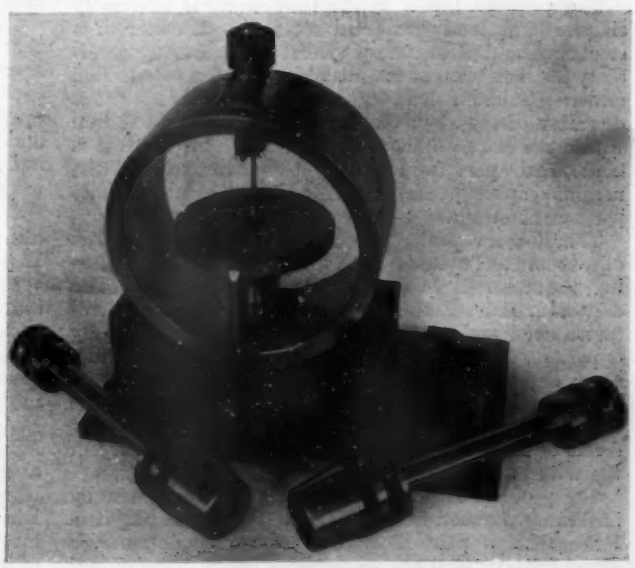


Fig. 1—A 10-ton capacity ring.

Unfortunately the compact design of modern tensile machines does not lend itself readily to the stacking or suspension of large quantities of weights and often only a fraction of the total load range is covered. The range may be increased by the use of standardised proving levers, but here again levers of more than 20 tons capacity are something of a rarity.

The Calibrated Test Bar

The calibrated test bar is the most simple of all the elastic calibration devices, and while it is unsuitable for formal verification purposes, it is extremely useful for routine checking. The cross-sectional area and tensile strength of the test bar should be such that when it is loaded to the maximum load capacity of the machine, it is only stressed to within roughly 75% of the elastic limit. It is a good plan to determine the load/elongation characteristics as accurately as possible immediately after the machine has been calibrated.

The bar should have integral shackles and spherical seatings to ensure axial alignment. Whenever it is desired to carry out a routine check test on a tensile machine, a film of oil is applied to the seatings and the bar mounted in the machine. Using a 6-in. Lamb extensometer, the extension is recorded for each ton increment of load from 1–20 tons.

A load-elongation diagram is then plotted and compared with the diagram constructed from the data obtained at the time the machine was calibrated. In this manner it is possible to detect surprisingly small discrepancies between the indicated loads.

The Steel Proving Ring

The problem of load capacity seldom arises with proving-rings for they are made in various sizes to accommodate loads from 10–50 tons. Although not so accurate as dead loading, the complete load range of the machine can usually be checked. They are easily transported (a 50-ton capacity ring weighs only 80 lb.) and they may be used on both vertical and horizontal machines in either tension or compression. Moreover the task of calibration takes much less time.

The principle upon which they work is extremely simple. The load is applied along the diameter of a steel ring, and means are provided for measuring accurately the deflection at the intrados. Since the deflection under maximum load is small compared with the diameter of the ring, the stress at any section is well within the elastic limit of the material from which it is made, and apart from loading it well in excess of the nominal capacity there is little danger of producing any permanent strain.

The 10-ton capacity ring illustrated (Fig. 1) consists of a short steel cylinder with integral screws for attaching the tension shackles or compression caps. The deflection under load, is measured by means of a micrometer screw and a vibrating reed mounted diametrically opposite. The reed is released after being displaced sideways, and while it is oscillating freely the micrometer is brought slowly towards it until a low musical note occurs on contact. This operation must be performed carefully for it is important that the duration and intensity of the note be maintained for all successive readings.

The problem of extraneous noise can be overcome by insulating the reed from its holder, and connecting it in series with a pair of high resistance headphones, a 4-volt battery, and the micrometer frame. The circuit is then momentarily completed each time the end of the reed brushes against the top of the micrometer screw.

The screw has 50 threads per inch, and there are 200 divisions on the micrometer disc so that each division represents 0·0001 in. When taking readings each division is subdivided into tenths by estimation, and each subdivision corresponds to a load of approximately 3 lb. on the ring illustrated.

Since the proving-ring and the tensile testing machine should be at very nearly the same temperature, the ring should be placed near or in the machine at least half an hour before calibration is commenced. The ring should then be subjected to a preliminary cyclic loading in axial tension or compression. All the load is then removed and a reading taken on the micrometer. After a satisfactory reading has been obtained the ring is sealed, and micrometer readings of the deflection are noted at suitable increments of load up to the capacity load.

The operator is provided with tables or a simple graph showing deflection in micrometer disc units plotted against calibration factors. To determine the true load in pounds at any indicated load on the machine, he has only to read off the appropriate calibration factor and multiply it by the number of deflection units.

Since the modulus of elasticity in flexure diminishes as the temperature increases, a correction must be applied to the readings for any marked deviation from normal room temperature at the time of calibration. Given that the ring itself was calibrated at a temperature of 70° F., and that the modulus of elasticity diminishes by 0·015% for every 1° F. increase in temperature, the correction may be obtained from the following formula:

$$\text{Ft.} = \text{F70}\ (1 - 0{\cdot}00015\ [t - 70]).$$

Where t = temperature at time of calibration.
Ft. = calibration factor at temperature.
F70 = calibration factor at 70° F.

The need for introducing calibration factors arises from the fact that the deflection of the ring is only approximately proportional to the load. When the formal calibration of the ring itself is being carried out, the load in pounds is divided by the deflection in micrometer disc units and the quotient is called the calibration factor for that particular load.

In actual practice it is customary merely to verify that the testing machine is within the specified limits of accuracy, rather than to determine the actual error at each particular load checked throughout the range. The operator has only to know the permissable maximum and minimum number of deflection units for each indicated load. In this manner it is comparatively easy to check a 50-ton capacity machine in two or three hours.

MR. N. E. KEARLEY, A.M.I.E.E., has been appointed Deputy Director to the Council of Industrial Design. With a wide practical experience of the engineering industry in various technical capacities, Mr. Kearley became Civilian Technical Officer at the Royal Aircraft Establishment. In 1929 he joined the Examining Staff of the Patents Office, and later served for several years in the Industries and Manufacturers' Branch of the Board of Trade. In 1939 he joined the staff of the Ministry of Supply, but in 1940 he was given leave of absence to join the Indian Army Ordnance Corps, in which he served successfully as Deputy Assistant Director of Mechanisation, as officer commanding the mechanisation design establishment, and as Chairman of the Tank Development Board, Calcutta. He returned to the Ministry of Supply in October, 1945.

The Structure of Nickel-Aluminium Magnet Alloys and the Control of Brittleness

By Alan Torry, A.I.M.

(*Research Department, Darwins Limited.*)

Following a historical survey of the progress in explaining the cause of high coercivity in the ternary alloys of Ni, Fe, and Al, an attempt is made to show how the factors responsible for coercivity are also associated with a weak intergranular structure. It is suggested that excessive brittleness in these alloys may be directly attributed to the differences in the rates of contraction or expansion of the phases already present or being precipitated. A study of the thermal dilatation of Alni, Alnico and Alcomax alloys of varying compositions revealed the magnitude of the influence of such elements as Al, Cu, and Ti, in causing departure from uniform expansion or contraction, both after casting and in subsequent heat-treatment. By controlling the relative proportions of all the elements present with improved knowledge of their separate effects on the magnetic characteristics, heat-treatment and dilatation behaviour, economic production has been improved and mechanical stability has been obtained without sacrifice of magnetic performance.

PERMANENT-MAGNETS made from nickel-aluminium-iron base alloys are of comparatively recent origin. It was in 1932 that the Japanese scientist Mishima, first called attention to their possibilities by introducing aluminium into the more familiar binary alloys. Since then development has proceeded apace and the original "Alni" alloy has been far outdistanced in its potentialities as a permanent magnet material by later discoveries involving the addition of other elements to the basic chemical composition. Chief among these elements are cobalt and copper, with titanium playing an important but secondary role.

Previous to 1932, the highest values of magnetic energy per unit volume of magnet material were obtained from the well-known 35% cobalt magnet steel (also discovered by a Japanese—Honda). The new Mishima "Alni" alloy soon showed itself capable of an average increase of about 25% in maximum useful energy above the 35% cobalt steel. The usual values obtained from 35% cobalt steel are about 38,000 ergs and from Alni about 50,000 ergs per cubic centimetre.

A typical example of "Alni" composition is 26% nickel 12% aluminium, the balance being practically pure iron. The alloy is maintained as pure as possible especially in that the carbon content from whatever source should be almost zero. (This stipulation regarding carbon applies equally to the later alloys now known by the general terms "Alnico" and "Alcomax," or "Ticonal.")

"Alnico" resulted from the substitution of part of the original nickel content by a nearly equivalent amount of cobalt, together with the addition of a low percentage of copper. A typical composition of "Alnico" is now:—

Ni .. 17%	Co .. 12%		Balance .. Fe	
Al .. 9%	Cu .. 3–7%			

For certain technical reasons titanium is also added in amounts under 2% by several makers.

Maximum energy values from such an Alnico alloy are approximately 70,000 ergs. per cubic centimetre.

In July 1938, an article in *Nature* described an experiment whereby an alloy of the Alnico type was heat-treated by cooling it from 1,200° C. in a powerful magnetic field. This had the effect of bringing about a further 20% increase in energy value in the cold sample, when measured in the same direction as the previously applied magnetic field. By late 1939, a provisional patent had been taken out covering the application of a powerful magnetic field to a new type of alloy with considerably higher cobalt content, whilst the alloy cooled at prescribed rates through the Curie point[1]. The improvement in energy values resulting from this combination of modified composition and anisotropic treatment was found to be startling. From the previous best performance of 70,000 ergs. per cubic centimetre the new figure rocketted to about 170,000 ergs., which remains the practical optimum value to-day.

Alloys which lend themselves to anisotropic treatment in this manner are found within the general composition.

Ni .. 11–15%	Al .. 7–10%		Balance .. Fe	
Co .. 14–30%	Cu .. 3–6%			

Thermal Equilibrium

The early work of Köster[2] on the thermal equilibrium of the ternary alloys of Ni, Al and Fe covered a much wider range of composition than that found to contain alloys of useful properties as permanent magnets. Köster's work did much to confirm that these alloys were of the precipitation-hardening type, and that they were derived from the existence of a miscibility gap in the ternary system. Köster postulated a two-phase area of α and β solid solutions extending through the system thus separating the α and β fields (Fig. 1A).

Precipitation-hardening was facilitated owing to the capacity of the compound Ni_3Al to form a series of continuous solid solutions with the β phase, and by the bulging of the saturation surface of this β solid solution towards lower Ni and higher Al contents with fall in temperature (Fig. 1B).

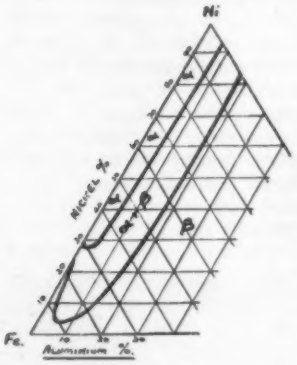
Supersaturation could thus be brought about at the lower Ni side of the miscibility gap over a fairly wide range of Ni content and, as Mishima pointed out, the normal rates of cooling from casting were generally favourable in bringing about those degrees of precipitation which assisted the development of useful magnetic coercive force in these alloys.

1 British Patent Specification No. 522-731.
2 W. Köster. "Das System Eisen-Nickel-Aluminium." *Arch. Eisenhüttenwesen* 1933-4, Vol. 7, p. 257.

Köster's work was strongly supported by results of magnetic tests on several series of alloys including the following :—

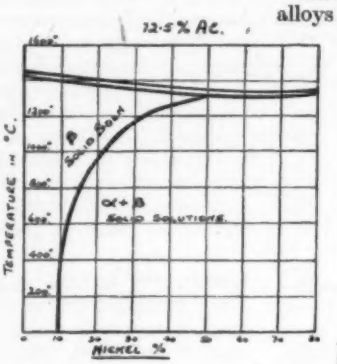
(a)	Ni constant at 15%		Al 6–20%	..	Balance	Fe	
(b)	"	"	20%	Al 6–20%	..	"	"
(c)	"	"	26%	Al 8–18%	..	"	"
(d)	"	"	33%	Al 8–20%	..	"	"
(e)	"	"	38%	Al 8–20%	..	"	"

The magnetic properties of the low Ni lower Al alloys are characterised by relatively high values of remanence and low values of coercivity.



(a)

Fig. 1a.—Portion of Fe-Ni-Al system at room temperature.



(b)

Fig. 1b.—Section across the thermal equilibrium diagram at 12.5% Al. (Both after Koster).

quoted have been found in the neighbourhood of Ni 26, Al 12, balance Fe.

Typical values obtainable from this type of alloy after heat treatment are Brem. 6,500, BHmax. 1·3 millions and Hc. 450. It will be seen that the " fulness-factor " relating to maximum energy has been improved rather than impaired by the compromise between the end values of the demagnetising curve (Fig. 2).

Micro-Structure

Although the explanation has been offered that the alloys so far described are fundamentally of the two-

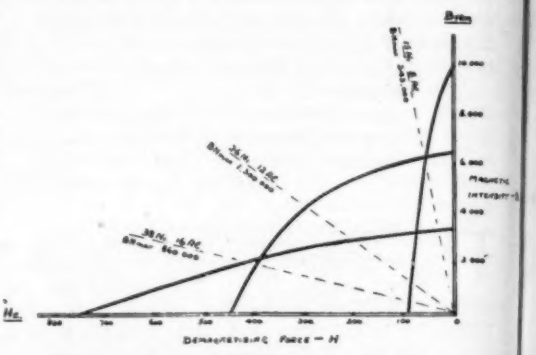


Fig. 2.—De-magnetisation curves of heat-treated " Alni " type alloys showing effect of composition.

The composition 15 Ni 8 Al, when air-blast cooled from 1,200° C. and tempered at 750° C. gave remanence (Brem) 9,900 Gauss. The coercivity (Hc) was 90 oersted and the maximum useful energy per cubic centimetre amounted to $\frac{343{,}000}{8\pi}$ ergs. ie. $\left(\frac{\text{BHmax.}}{8\pi}\right)$

Increasing aluminium content in series (*a*) up to 12% lowered the initial solution temperature required to 800°C. Above 12% aluminium the composition is evidently outside the range of the miscibility gap since the alloys possess a'most no coercivity—both " as cast " and after heat-treatment from as high as 1300°C.

Increasing the nickel content to about 30% accentuates the influence of the α phase so that a corresponding increase in the aluminium content is required to maintain a stable two-phase structure. With 20% nickel the range of miscibility is increased to about 15% Al. As the nickel content rises to 26% the highest aluminium content to fall within the miscibility gap rises still further to about 20% which appears to be the maximum for all values of nickel.

As compared with the low Ni, low Al types, alloys of the approximate composition Ni 38, Al 16, balance Fe are then characterised by low remanence and high coercivity values of the order Brem. 3,700, BHmax. 860,000, Hc. 750, after quenching from about 1,250° C. and tempering at 700° C.

The usefulness of any alloy as a permanent magnet depends on a suitable ratio of its remanence and its coercivity, quite as much as its maximum energy per cubic centimetre, if the magnet's section to length ratio is to be adjusted to conform with convenience in design. Alloys which effect a compromise of this kind between the remanence and coercivity values already

phase type, very little is discernible of a second phase in the metallographic inspection of many of them apart from a general discolouration of the main solid solution or occasional traces of a second constituent in the grain boundaries. Such is the normal appearance of the majority of the magnetically useful alloys of this type when examined immediately after casting or quenching treatment (Fig. 3). Only after prolonged annealing at high temperature is there any substantial indication of precipitation throughout the original solid solution (Fig. 4). In this more or less fully precipitated condition the alloys do not possess useful magnetic property but the intrinsic hardness is reduced, in certain cases, sufficiently to permit limited machining as distinct from grinding.

There is frequently a considerable improvement in maximum energy to be derived from tempering following quenching but this phenomenon is not accompanied by any noticeable change in the micro-structure.

Later Conceptions

X-ray examination subsequently showed that the two-phase area was more complex than at first supposed. Bradley and Taylor[3] suggested that Köster's two-phase area consisted of four separate fields instead of one. The miscibility gap could be divided into $\alpha + \beta$, $\alpha_1 + \beta_2$, $\alpha + \beta + \beta_2$ and $\beta + \beta_2$ regions where according to their own convention α is face-centred cubic and β is body-centred cubic in atomic arrangement (Fig. 5).*

3 A. J. Bradley and A. Taylor. " An X-ray investigation of the cause of high coercivity in Iron-Nickel-Aluminium Alloys." *Nature*, 1937, Vol. 140, p. 1,012.

* The suffixes (1) and (2) indicate the presence of superlattices and " m " and " n " distinguish the magnetic and non-magnetic regions at room temperature.

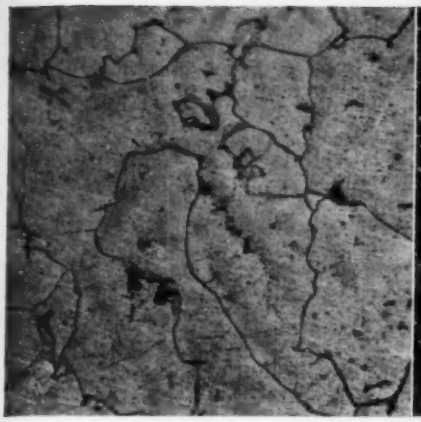


Fig. 3.—Photomicrograph of typical "Alni" alloy slightly etched (× 100).

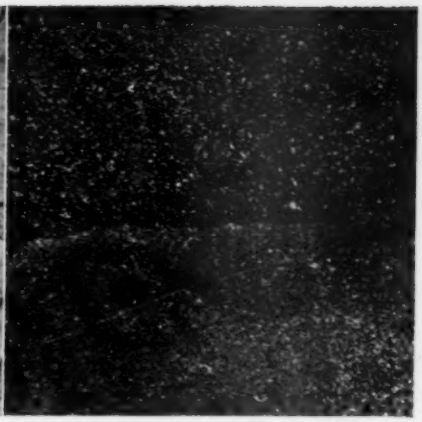


Fig. 4.—Photomicrograph of annealed "Alni" alloy etched ferric chloride (× 200).

Single-phase areas in the diagram have curved boundaries whereas two-phase areas are covered by "tie lines" whose extremities joint the boundaries of the distinct phases. Similarly, 3-phase areas are triangles. "An alloy of composition represented by a point in a two-phase area breaks up into two phases of compositions represented by the ends of the tie line running through the point from one single phase area to another." Also "an alloy anywhere within a three-phase area breaks up into three phases of the compositions represented by the vertices."[4]

Following this hypothesis "Alni" alloys approximating to the composition Fe_2NiAl will tend to decompose from β_2 solid solution into β and β_2 if allowed to cool slowly enough from solution temperature. Both these phases are ferro-magnetic, but there is a very narrow margin of difference between their lattice dimensions and consequently in the process of separating out in cooling, the atoms of the iron-rich β-phase are held apart under a condition of internal stress, unless the rate of cooling be sufficiently slow to permit disentanglement of the atomic forces of the separate phases. Under these conditions—equivalent to prolonged annealing — the separate phases can be resolved by micro-inspection at high power.

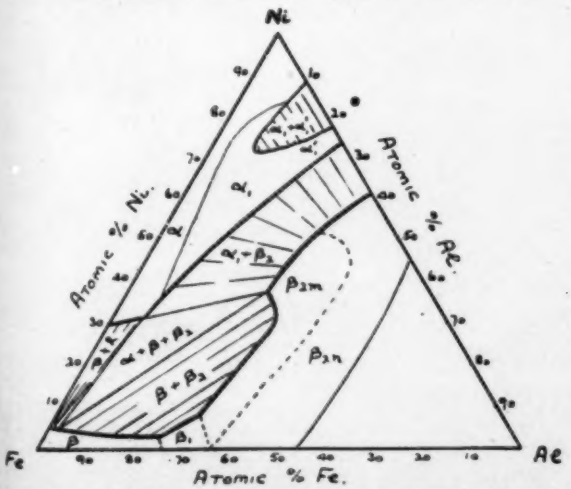


Fig. 5.—The system Fe-Ni-Al at room temperature based on X-ray work by Bradley and Taylor.

With the rates of cooling obtained by casting or heat treatment for the development of magnetic hysteresis the two phases although present are not in a state of visible separation but are in fact atomically intermingled. The resultant internal strain of the atoms of the β-phase is the prime cause of the high coercivity.

The later addition of cobalt, copper and titanium, into the composition of the basic Ni-Fe-Al apparently does not introduce further considerations of a metallographic nature into the system. These elements do, however, profoundly influence the degree and conditions of maximum development of hysteresis in the alloys. One of the earliest observed effects of the addition of cobalt, copper and titanium in the later Alnico was that their addition made slower rates of cooling permissible and particularly in the case of titanium, lower solution temperatures effective. Titanium also made a special contribution in reducing the grain size considerably.

Commercial Production

The commercial development of these alloys has been based hitherto on the most economical values of maximum energy retained by any one of them within a fairly wide range of composition. This approach to the matter took little or no account of the technical difficulties which might arise. Incidentally, there has always been an important limiting factor in their commercial application, namely, that it has not been found possible to forge them or to prepare forgeable varieties.

This is understandable on the grounds that the introduction of aluminium to the basic Ni-Fe series induces the formation of intermetallic compounds in the mutually soluble nickel and iron. Intermetallic compounds are also formed by aluminium in the presence of cobalt and copper which are added to the later alloys, Alnico, etc. Such compounds are notoriously hard and brittle and since aluminium appears to be an essential ingredient there would seem to be little prospect of forging qualities being developed.

Heat Treatment

In general, the heat treatment found necessary to promote the most favourable degree of super-saturation and precipitation in both Alni and Alnico alloys consisted of raising the castings to between 1,000° C. and 1,300° C. and then quenching, followed in most cases by a subsequent tempering treatment between 600° C. and 700° C.

4 A. J. Bradley and A. Taylor. "Investigation of Equilibrium Diagrams of Ternary Alloys by X-rays." *Nature*, 1937, Vol. 140, p. 543

The nature of the quench was largely determined by the mass of the individual casting but also had reference to the precise composition of the alloy. Cooling in still air, air blast, and oil are the three standard means of quench commonly available. Depending on the size of casting these agencies are capable of cooling rates between 25° C. and 200° C. per minute, within the range of temperature from solution treatment to approximately 600° C. for most magnet work. On reaching 600° C. castings were allowed to go cold naturally or transferred for tempering.

With certain compositions annealing was done successfully by heating to 1,000° C., soaking thoroughly and then cooling very slowly at a maximum rate of 30° C. per hour down to below 600° C. Although hardness values of up to 450 Vickers resulted from this treatment it was found possible to drill these materials at commercial speeds with high-speed steel drills. Milling, tapping, turning and even sawing occasionally have been successfully accomplished.

Granular Structure

Nickel-aluminium-base magnet alloys are characterised by structures of relatively large grain size. As compared with steels and most other alloys they are inherently brittle and require careful handling if excessive breakage is to be avoided. This is not by itself a very serious disadvantage since most applications of permanent magnets do not demand a high degree of mechanical strength or resistance to shock. On the other hand there are numerous applications in which it would be a very serious matter should any crumbling of the magnet result in dislodged particles fouling the mechanism of the apparatus in which the magnet is incorporated. Examples of such uses are ammeters, voltmeters, galvanometers, small generators, etc. Thus quite apart from the necessity of inducing the maximum degree of useful magnetic energy in any particular alloy it is highly desirable that the alloy should also possess reasonable mechanical stability.

In the early period of commercial manufacture of magnets of these types considerable loss of production was all too frequent from cracking and crumbling as the magnets passed through the various stages of fabrication. The most serious troubles arose, however, during the heat treatment process, but final surface grinding accounted also for a high proportion of failures.

Analysis of production records showed that certain alloys were more prone to cracking than others and that individual patterns of magnet tended to crack in well-defined locations of the casting, such as re-entrant corners. Other cracks were liable to occur from any surface irregularity such as a small casting lap or fold, whilst one very puzzling and persistent location was on the internal surface of closed ring castings, or inside cored holes.

Thermal Dilatation

Accordingly a study was initiated of the thermal dilatation of nickel-aluminium base magnet alloys in general, especially with regard to the influence of apparently small variations in composition.

The first step in these experiments was the determination of any thermal change points or "arrests" there might be during the heating and cooling periods. With these alloys the "arrests" were found to be not of the well-defined point type but extended over a wide range of temperature both in heating and cooling, and appeared to indicate absorptions and releases of heat of considerable magnitude. Since these thermal effects were utilised in promoting metallurgical changes it was desired to measure their effect on the volume of the alloys as they were being experienced.

A special form of dilatometer was constructed in conjunction with a furnace capable of heating a test sample uniformly to just over 1,300° C. The dilatometer functioned on the basis of the difference in expansion

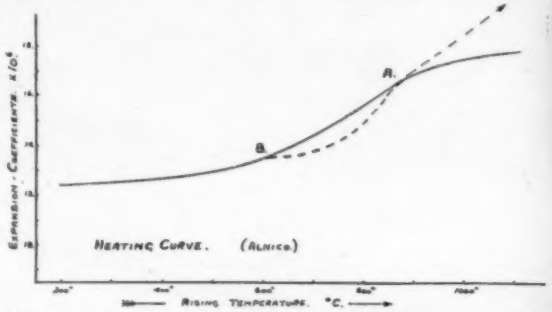


Fig. 6.—Thermal expansion curve obtained by heating typical "Alnico" alloy. Full line—previously annealed. Broken line B— A— as cast.

and contraction between the test sample and a pure silica comparator rod. Silica rod extensions outside the furnace caused small mirrors to tilt over, thus reflecting into telescopes the image of a graduated scale at a fixed distance from the mirrors. One mirror accounted for differences in the lengths of the test-sample and silica comparator rod. The other mirror gave any corrections necessary due to the unpredictable movement of the whole furnace system.

Fig. 7a.—Derived curve of possible internal strain on cooling.

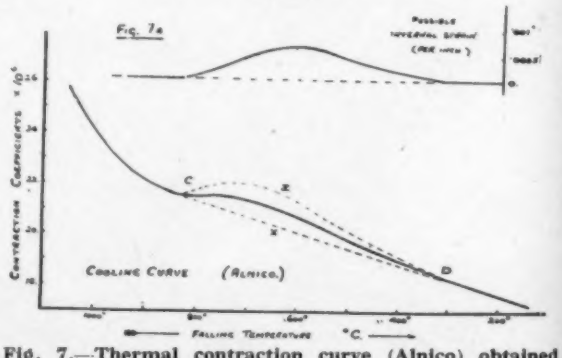


Fig. 7.—Thermal contraction curve (Alnico) obtained from slowly cooled alloy.

It was not considered practical at first to work at any higher temperature than 1,100° C. owing to the probability of the silica comparator and extension rods becoming soft or plastic giving erroneous results. The test sample and comparator rods were encased in a thick cylinder of pure nickel so as to maintain them at a steady uniform temperature throughout their whole lengths.

The method of recording and calculating results requires a word of explanation. Results were plotted with reference to the initial steady temperature both

Fig. 8.—Photomicrograph of Alnico alloy as cast, deeply etched (× 100).

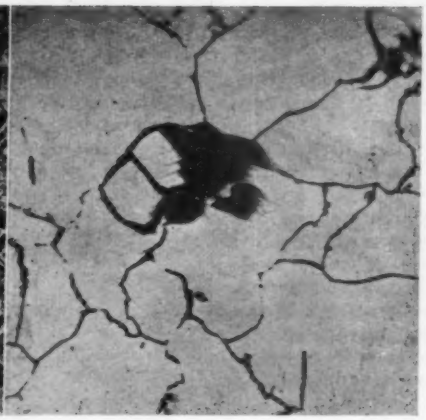


Fig. 9.—Photomicrograph of Alnico alloy annealed, unetched, showing loose grains and open structure.

on heating and on cooling. Thus the coefficients obtained on heating are referred to as "expansion coefficients" and those on cooling as "contraction coefficients" (Figs. 6 and 7). It will be realised that on this basis even if the dilatation recorded were entirely reversible the numerical values of expansion and contraction coefficients would be different, but it was considered that this method gave a more direct picture of the behaviour of the alloys.

The main feature of the dilatation curves obtained in this manner is their departure from the smooth arc contour governed by the formula $\frac{L-La}{La} = \alpha T + \beta T^2$ which holds good during the maintenance of a stable physical condition within the specimen. This departure takes the form of a more or less pronounced "hump" in the curves.

From a study of the significance of this hump and its manner of development in a large number of Alni, Alnico and Alcomax type alloys some very useful inferences have been drawn. These inferences have been very largely substantiated by their subsequent application in the commercial production of these alloys.

On heating, the departure from uniform dilatation of annealed samples no doubt coincides with the onset of solution of the precipitated phase (Point A in Fig. 6). The heating curves of samples in the "as cast" condition, however, frequently show an earlier inflexion at much lower temperatures (Point B in Fig. 6). This is attributed to the resumption of precipitation which was in progress whilst the samples cooled from the molten state, but suppressed when the temperature drop no longer permitted the requisite molecular activity.

On cooling, the deviation of the contraction curve from a smooth arc will no doubt coincide also with the beginning of precipitation of the β phase out of the parent solid solution (Fig. 7, Point C). From this point onwards to room temperature the resultant curve will be an average of the contraction curves of the two or more phases present and will vary with their degree of precipitation. Extrapolation of the original curve to room temperature will give a fairly reliable contraction rate of the remaining solid solution (dotted line X). It will be seen that the extrapolated portion of the curve later approaches the plotted curve so closely and so much in parallel that for all practical purposes it may be regarded as rejoining it at some point such as D in Fig. 7.

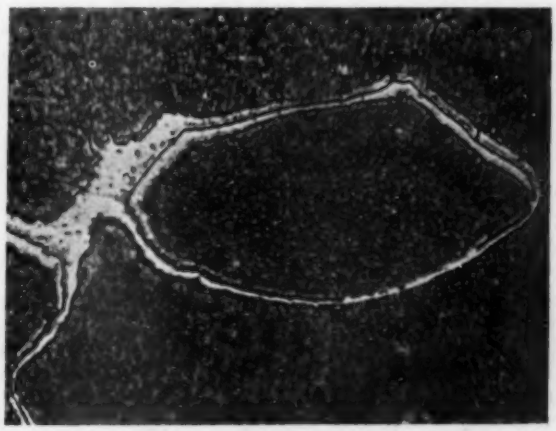
From one extreme and the known average the opposite extreme can be plotted also (dotted line Z) giving rise to an area which may represent the development of strain within the alloy. Whether or not this development will cause cracks to occur in the alloy depends on the plasticity or capacity to absorb the strain at any given temperature.

When an alloy cools at quicker rates which tend to restrict complete precipitation (as under normal heat treatment) a state of super-saturation develops which increases towards room temperature.

Now the relatively large grains composing the structure are derived from the cast state of the alloy and the solid solution of which they are formed is not likely to be fully homogeneous (Fig. 8—showing "cored" structure). Consequently the regions of the grain boundaries are probably richer in the lower freezing point elements such as aluminium and copper.

Super-saturation of the solid solution by the iron-rich β-phase will develop earliest therefore in the heart of the grain, causing lattice distortion. When precipitation occurs it will tend to bring about contraction of the individual grains away from each other placing a stress on the grain boundaries, thus weakening them and giving rise to a loosely held or brittle structure (Fig. 9 showing loose grains).

Fig. 10.—Photomicrograph of annealed Alnico, showing alumina segregation in grain boundaries. Note cleavage of alumina as grains receded, etched ferric chloride (× 500).



The difference between the resultant lengths of the precipitated and unprecipitated phases on cooling from solution temperatures are shown in the derived curve of possible internal strain (Fig. 7A). This offers an explanation of the occurrence of cracks on inner surfaces of closed rings and cored holes in magnet castings. Precipitation will be more likely to occur in portions of a casting which are subjected to slower rates of cooling. Due to relief from lattice distortion such regions will contract differentially as compared with the more rapidly cooled regions still held distorted, as required for the development of coercivity.

Another phenomenon explained by this hypothesis is the presence of temper-coloured fractures in the heart of castings broken open after heat-treatment. The discoloured surfaces are often completely surrounded by perfectly sound material but usually are associated with small shrinkage cavities. The air trapped in these cavities would be responsible for the slight oxidation of the cracks developed later.

In all these alloys aluminium is the most difficult element to control owing to its high rates of oxidation whilst melting. It had been noticed that when the aluminium content tended to be higher than nominal specification, the magnets produced were more brittle than usual.

Excessive oxidation of aluminium may be caused when it is added to the melt, owing to its much lower specific gravity which encourages it to float on top in contact with the atmosphere. When this happens any alumina not removed with the slag may be distributed ultimately in the grain boundaries adding to their weakness (Fig. 10). It is important therefore, to ensure that the aluminium is kept submerged in the melt until completely dissolved.

Dilatation curves from a series of Alnico type alloys with Ni 17, Co 11, Cu 6·5 and Al ranging from 8% to 13% illustrate the intensification of this condition of strain with increase of aluminium up to 12% (Fig. 11).

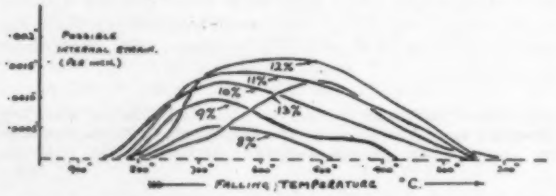


Fig. 11.—Derived curves showing effect of aluminium content on liability of Alnico to develop cracks on cooling from 1,100° C.

Increase in aluminium dilutes the concentration of the iron-rich β-phase in the solid solution progressively so that super-saturation reaches its peak at lower temperatures, and when this is relieved as strain the temperature may be low enough to prohibit plastic flow, with resultant cracking more probable in the higher aluminium alloys.

At 12% Al this kind of strain is at a maximum. Probably this is due to the close proximity of the upper Al limit of the miscibility gap for these alloys. Similar alloys with higher Al content become useless from a magnetic point of view and suffer from a different kind of brittleness. This fragility is not so strongly intergranular but resembles more the brittleness of pot in the dull amorphous appearance of the fracture, as compared with the brilliant lustre of the magnetic alloys.

The function of copper and titanium additions is apparent from magnetic results which show intensification in the development of coercivity. Since coercivity is associated with internal stress it is not surprising to see the great difference in possible strain developed on cooling between similar alloys where copper or titanium

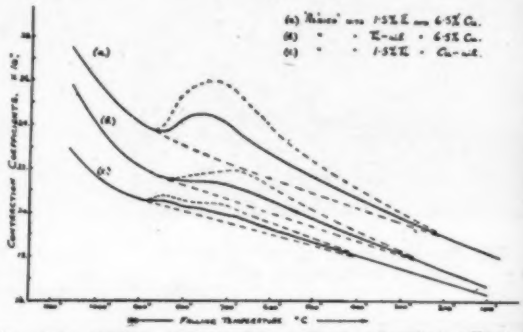


Fig. 12.—Alnico contraction curves showing effect of copper and titanium on cracking tendency.

were present in the one and absent in the other. (Fig. 12).

Whilst being very useful in assisting coercivity values copper can be a dangerous source of cracking if added indiscriminately. On the other hand its ability to encourage precipitation sometimes outweighs this consideration. This is particularly the case where ease of machinability is desired in the annealed condition. Consequently one machinable variety of Alnico contains 6·5% copper as against a more crack-proof but unmachinable composition with 3·5% copper plus 1·0% titanium. The strong influence of copper to increase liability to cracking has been shown to exist not only with the "Alnico" type alloys but also in connection with the earlier Alni and more recent Alcomax alloys.

Titanium, however, has two especially redeeming features inasmuch as it promotes considerable grain refinement in the structure with the very smallest additions of about $\frac{1}{4}$% and also that solution temperatures required are reduced. Grain refinement ensures the provision of a larger number of grains per unit volume, ostensibly by the creation of a larger number of nuclei in the solidifying metal. Hence solidification is accompanied by a reduced degree of segregation of low freezing point constituents in the grain boundaries. Consequently, the degree of tension in the grain boundaries resulting from precipitation contraction is reduced. A further result of this effect is to improve machinability of annealed alloys owing to the restricted development of hard constituents in the aluminium-rich matrix.

It is obvious from magnetic results that re-solution of the phases present as cast is a progressive phenomenon under high temperature treatment. Where there is a wide divergence of chemical composition inside individual grains, heat treatment temperatures will be relatively high. In producing greater homogeneity titanium additions result in lower solution temperatures, which is a factor of considerable commercial importance.

The first demonstration of this fact was due to Linley[5] who was able to make commercial use of Alni

5 British Patent Specification No. 446,894.

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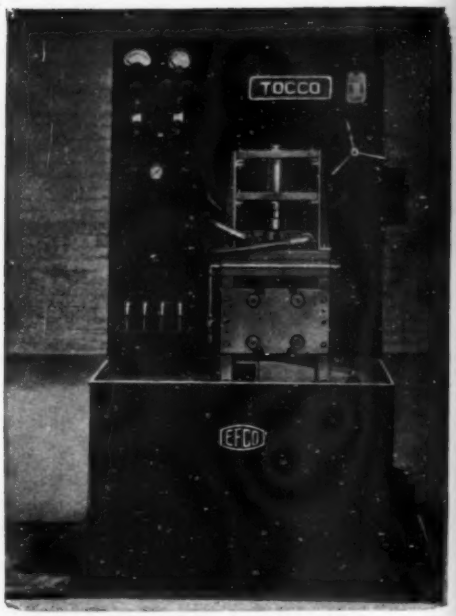

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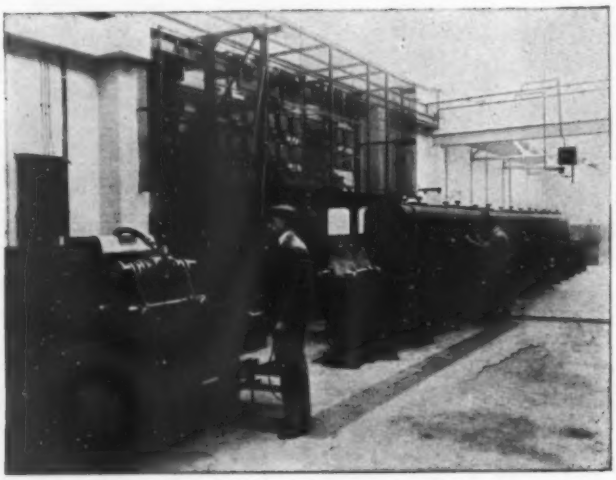

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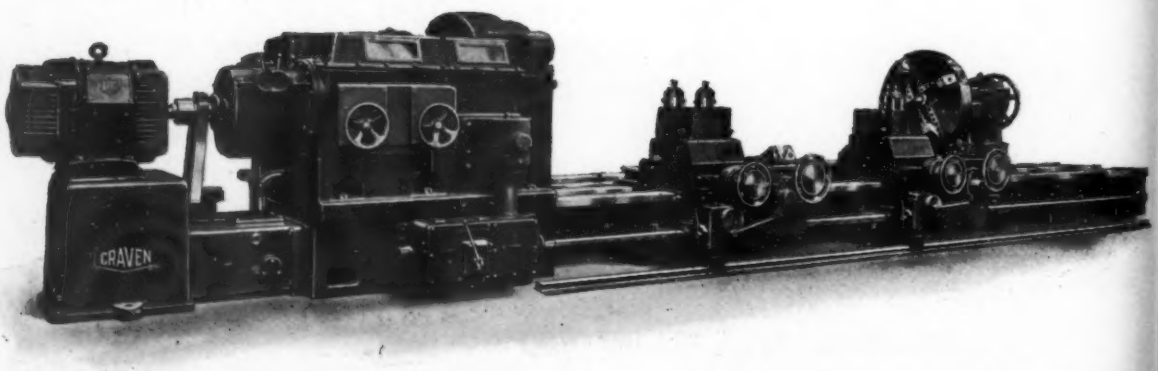

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alloys outside the range of the Mishima patent with over 30% nickel owing to his addition of titanium in small quantities. Without titanium these alloys require a solution treatment of over 1,300° C. bordering on the liquidus. With 1% or so titanium added to the melt heat-treatment became an industrial proposition at about 1,200° C. to 1,250° C.

Anisotropic Alloys

With the advent of the high cobalt anisotropic alloys the picture already presented becomes still further complicated. Present acquaintance with them indicates that conditions favourable to optimum magnetic potentiality are established at even higher temperatures and the rates of cooling through these temperatures rather more critical than with the Alnico alloys. Dilatation curves of a 14% Ni, 8·5% Al alloy plotted from 1,330° C. downwards have revealed that precipitation of the kind so far referred to begins about 1,275° C. in the very slowly cooled alloy, but is largely suppressed by normal rates of cooling.

Another alloy with under 12% Ni gave a similar point at about 1,175° C. These temperatures are far in excess of any ferromagnetic condition so that the later orientation in a magnetic field is seen to be a secondary phenomenon. Jellinghaus[6], who also used dilatometer methods in his investigation of these anisotropic alloys, confirms the absence of transformations in the temperature range 20°–1,000° C. Nevertheless he agrees that they are heterogeneous in structure.

Early attempts at heat treatment of the alloys with 14% Ni employed temperatures of over 1,300° C. Later it was found that such high temperatures were not always necessary, depending on the favourable condition or otherwise, of the alloys as resulting from casting.

Here again, titanium has proved useful in reducing the initial temperature of magnetic heat treatment of samples annealed at 1,000° C. for machining. Where annealing is not required, the titanium appears to standardise the fundamental condition of atomic stress in the cast material so that anisotropic properties may be superimposed by heating to a maximum temperature of 950° C. and then cooling in a magnetic field.

It will be readily appreciated that magnets whose structure has not been further loosened by exposure to very high temperatures are likely to be much stronger mechanically.

Acknowledgment

The author wishes to thank the management of Messrs. Darwins Ltd. for permission to publish this paper.

6 W. Jellinghaus. "Iron-nickel-aluminium-cobalt-copper alloys with preferred magnetic orientation." *Archiv. fur das Eisenhutten-wesen*, 1943, Vol. 16, p. 247–251.

The Ellipsoid Melting Furnace

By A. J. G. Smith

Of the non-crucible melting furnaces probably the best known in this country is the horizontal rolling or rocking furnace. A modification of this type has been in use on the continent in recent years which is mounted at an angle of 30°. Brief particulars are given of this furnace, together with some typical melting performances.

IN the past some noteworthy melting results have been achieved in non-crucible furnaces in which the flame from the burner is admitted direct into the melting chamber. Of these types, probably the horizontal rolling furnace is the best known as it is manufactured and has been considerably developed in this country. Of recent years, however, a modification of this type has begun to be used in certain overseas countries which differs from its prototype in certain important features. This new development which is known as the Ellipsoid furnace is shown by the sectional view reproduced in Fig. 1.

It will be seen that in general construction it consists essentially of a strong sheet-metal outer casing of the drum type, which instead of being mounted horizontally on two trunnions is permanently tilted to an angle of some 30° from the horizontal. Of the two points of support necessary, one consists of a pin located in a forged end-piece riveted to the furnace foot-plate and mounted in a suitable ball bearing; the other point of support consists of a heavy wrought-iron ring riveted round the circular furnace body, which rests upon rollers which can be actuated by an electric motor, so that a rocking motion can be imparted to the furnace.

Oil firing is usually adopted for this type of furnace although of course, it is equally suitable for gas firing.

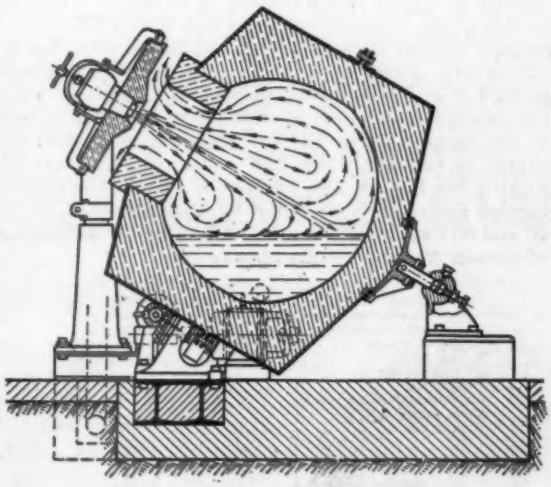

Fig. 1. Sectional Diagram of Furnace.

The burner is mounted upon a vertical pillar which also serves to carry the cover and operates under an air pressure of 1½ lb. per square inch. The furnace lining

is either of special shape firebricks or stamped, the latter method being generally recommended as being cheaper and having a longer working life equal to about 125 to 130 tons output before relining is required. Under working conditions, the swing cover is opened and the metal charged on to the hearth, after which the blast from the burner is directed against the opposite wall of the lining, the products of combustion being discharged through the vent holes surrounding the burner.

While the melting is proceeding, the driving rollers

Fig. 2. General View of Furnace.

are rotated by means of the reversing motor through suitable gearing in such a manner as to impart a rocking motion to the furnace body by which means the molten metal is constantly washed over the incandescent lining. The degree of rock given to the body can be regulated so that it is just short of allowing the metal to reach the tapping hole which is situated in the bottom part of the furnace below the jointing ring, but when the charge is brought down the motor which serves for rocking the furnace also effects the pour. The type of furnace described is now built in capacities of 6, 10, 20, 40 and 60 cwt. (brass), and the following are particulars of certain working results obtained :—

Metal	Red brass swarf	
Heating-up time	30 min.	
Oil consumption	5·6 gal.	
Weight of charge	lb.	668
Melting time	min.	37
Oil consumption	%	8·25
Pouring time	min.	18
Oil consumption for pouring	gal.	1·3
Metal poured	lb.	651
Melting loss	%	2·6

Metal	Red brass grinding swarf	
Heating-up time	30 min.	
Oil consumption	12½ gal.	
Weight of charge	lb.	552
Melting time	min.	30
Oil consumption	%	10
Pouring time	min.	10
Oil consumption for pouring	gal.	2·1
Metal poured	lb.	426
Melting loss	%	22·9

After the last-mentioned melt, the following two charges were run :—

Metal	Red brass scrap	German silver grinding swarf
Weight of charge (lb.)	902	697
Melting time (min.)	60	45
Oil consumption (%)	11·3	10·2
Pouring time (min.)	13	15
Oil consumption for pouring (gal.)	2	3·3
Metal poured (lb.)	660	613
Melting loss (%)	33·5	12

and the following are figures taken for the melt of another alloy, including oil consumption for heating up from cold :—

Metal	Argentan
Heating-up time	78 min.
Oil consumption	13·2 gal.
Weight of charge (lb.)	1392
Melting time (min.)	103
Oil consumption (%)	12·9
Pouring time (min.)	75
Oil consumption for pouring (gal.)	12
Metal poured (lb.)	1197
Oil content of swarf (%)	3·2
Melting loss (%)	11·5

Gold Output Rising in W.A.

GOLD production in Western Australia is mounting steadily. For the first five months of the current year 206,121 fine ozs. have been won compared with 185,402 ozs. for the corresponding period (January-May) of 1945.

In pre-war years annual production of more than one million ozs. was customary. Difficulties confronting the industry at present include a serious shortage of explosives, and while such conditions persist a return to pre-war levels of production cannot be expected.

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WESTERN AUSTRALIA GOLD PRODUCTION.

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January	36,708	41,491	42,140
February	33,176	35,867	37,271
March	38,550	38,764	39,533
April	26,757	35,078	41,107
May	37,708	34,202	46,071
June	40,523	36,591	—
July	36,546	39,236	—
August	60,186	59,414	—
September	39,443	33,464	—
October	36,998	34,109	—
November	35,771	37,760	—
December	42,074	37,760	—
	464,440	463,521	206,121

Beryllium Master Alloy Prices

IN view of the increasing interest in beryllium as an alloying element in copper, magnesium and aluminium, the announcement, by Beryllium Smelting Co., Ltd., of new prices for master alloys is of interest. This company can now supply ternary beryllium–magnesium-aluminium master alloy, with a nominal analysis of 5% beryllium, 5% magnesium, balance aluminium, at £9 per lb. of beryllium contained in the alloy, without charge for the magnesium or aluminium. The price for beryllium-aluminium master alloy, usually containing approximately 5% beryllium and 95% aluminium, is reduced to £9 per lb. of beryllium contained in the alloy with no charge for the aluminium. These prices are for lots of a minimum of 5 cwts., which are delivered free.

The price of beryllium-copper master alloy, containing approximately 4% beryllium, remains unchanged at £5 per lb. of beryllium plus 1/– per lb. of alloy, delivered free.

Control of Shrinkage Porosity and Mechanical Properties in Chill-Cast Phosphor-Bronzes, Leaded Bronzes and Gun Metals—II

By W. T. Pell-Walpole, B.Sc., Ph.D.
Gibbins Research Fellow, Metallurgy Department, University of Birmingham

The variables in the chill-casting process which affect the extent and distribution of shrinkage porosity have been examined for phosphorus bronzes, gun metals and leaded bronzes. The rate of pouring is the most important factor, and freedom from the most harmful types of shrinkage porosity is obtained when the rate is the lowest which will fill the mould without cold shuts. This critical rate is related to the shape and size of the ingot by the formula

$$R = KP$$

where R is the pouring rate, P is the periphery of the ingot cross-section (or in the case of cored sticks, the sum of the peripheries of mould and core), and K is a constant, which for constant mould conditions and pouring temperature, depends on the alloy composition. Values of K were determined for all standard bronze compositions.

Mechanical properties of these bronzes in general vary with casting conditions in the same manner as density—i.e., optimum values are obtained, uniformly across the section, when the pouring rate is the minimum which will fill the mould without cold shuts. Slight variations in properties are also produced by the effect of pouring and mould temperature on the distribution of the brittle compound phases.

Part II—The Effect of Mould Conditions

(*a*) *Mould Temperature.* Melts of 2B8 bronze (10% tin, 0·75% phosphorus) were prepared as before, and poured into the 1 in. dia. mould at 6 lb./min., and 1,200° or 1,250° C. The mould was dressed with tallow/graphite for all experiments, but mould temperature was varied from 20° to 200° C.

Visual examination of the ingots showed that for the range 20°–100° C. mould temperature had little effect on the quality of the surface or of a cross-sectional fracture. With higher mould temperatures, traces of central discolouration indicating shrinkage fissure porosity appeared in the fractures, and surface blowing porosity occurred more readily. Examination of microstructures confirmed the appearance of central shrinkage fissures with the highest mould temperature, and showed also a marked effect of mould temperature on cooling rate below the solidus in so far as this is reflected by the amount of coarseness and disposition of the ($\alpha+\delta$) eutectoid (see Figs. 26 and 27). As the mould temperature was increased, this constituent became coarser in resolution and less in quantity, and tended to occur in isolated areas, rather than as continuous intercrystalline envelopes.

Density measurements (Table II) showed little variation for mould temperatures of 20°–100° C., but an appreciably lower value was obtained for the highest mould temperature (pouring temperature 1,250° C.), confirming the observations of slight porosity in fracture and microstructure. Tensile values showed the same effect, but hardness decreased gradually with increase of mould temperature over the whole range. This was probably caused by the accompanying decrease in the amount of hard δ phase present, rather than by the slight increase in porosity.

The effects of mould temperature on the extent of shrinkage porosity are shown by these results to be very slight for the range examined—*i.e.*, 20°–200° C. It is clear that a further increase of mould temperature would cause an increase of shrinkage porosity and a greater tendency to blowing, together with increased difficulties of manipulation. It is concluded that moulds should be used at the lowest temperature consistent with freedom

TABLE II.

EFFECT OF MOULD TEMPERATURE ON THE PROPERTIES OF CHILL-CAST DE-GASSED PHOSPHOR BRONZE (2B8) INGOTS 14 IN. ×1 IN. DIA. POURED AT 6 LB./MINUTE.

Mould Temp. °C.	Pouring Temp. °C.	Analysis		Density. Grms./c.c.				Tensile Properties				D.P. Hardness	
				Cast		Machined		Top		Bottom			
		Sn. %	Phos. %	Top	Bottom	Top	Bottom	U.T.S. (Tons/sq. in.)	El. % on 2 in.	U.T.S. (Tons/sq. in.)	El. % on 2 in.	Centre	Edge
20	1,250	10·9	0·86	8·75	8·75	8·73	8·72	27·9	9	27·7	9	122	125
20	1,200	11·0	0·65	8·79	8·79	8·78	8·79	28·1	15	26·0	12	115	117
40	1,250	11·1	0·84	8·78	8·79	8·77	8·76	30·2	20	29·4	16	118	123
40	1,200	11·0	0·64	8·84	8·80	8·80	8·77	29·3	20	26·8	10	112	118
60	1,250	11·2	0·94	8·70	8·70	8·69	8·69	29·2	13	26·8	9	122	120
60	1,200	11·1	0·64	8·79	8·76	8·76	8·74	26·5	12	24·6	9	110	115
80	1,250	11·1	0·84	8·76	8·77	8·76	8·76	28·3	13	27·8	15	127	132
80	1,200	11·5	0·69	8·81	8·80	8·81	8·80	29·5	20	26·8	11	124	124
100	1,250	10·9	0·88	8·74	8·70	8·75	8·68	28·7	14	25·2	12	122	121
100	1,200	10·9	0·66	8·78	8·77	8·70	8·77	27·5	14	27·4	14	115	114
200	1,250	10·8	0·67	8·72	8·74	8·71	8·73	27·6	12	26·1	16	113	112
200	1,200	10·6	0·86	8·78	8·79	8·78	8·78	26·4	12	26·6	12	111	114

from the danger of deposition of surface moisture (*i.e.*, ∽ 100° C.).

In view of these results it was not considered necessary to repeat these tests with other compositions.

(*b*) *Mould Dressings.*—Mould dressings are normally selected with the object of obtaining good surface quality in chill cast ingots, but since they form an insulating layer between mould face and ingot surface, they may affect the cooling rate, hence the extent of shrinkage porosity in the ingot.

In order to restrict as far as possible the interfering effects of surface blowing (aluminous cement, french chalk), graphite rubbed into the mould surface, and a typical combustible type consisting of equal parts of tallow and powdered charcoal. The moulds were cleaned thoroughly between successive tests.

The surface quality of the ingots produced varied very considerably with the dressing used. Blowing occurred with both refractory dressings, and there was local surface roughness with the graphite dressing. The tallow/charcoal gave the smoothest and soundest surface.

Internal shrinkage porosity as shown by fractures, microstructures, density and tensile tests (Table III) showed no variation as between the refractory dressings and the graphite, but the volatile tallow/graphite caused a slight increase in central porosity. This was probably due to the insulating layer of gas produced between the mould face and the ingot surface, but it is possible that traces of the gas are dissolved and later ejected by the metal, causing slight gas porosity. It is not possible to distinguish between gas and shrinkage cavities when these are very minute and few in number as in these ingots.

TABLE III.

EFFECT OF MOULD DRESSINGS ON THE PROPERTIES OF CHILL-CAST INGOTS OF DE-GASSED PHOSPHOR BRONZE. POURED AT 0·14 LB./SEC., 1,250° C.

Mould Dressing	Analysis %		Density gm./cc.				Tensile Properties			
			Bottom		Top		Bottom, Cast		Top, Ann	
	Sn	Phos.	Cast	Mach.	Mach.	Ann.	U.T.S. Tons/sq.in.	El. % on 2 in.	U.T.S. Tons/sq.in.	El. % on 2in.
Blacklead	9·1	0·06	8·84	8·83	8·84	8·82	22·0	19	22·0	47
50/50 Tallow/Charcoal	8·9	0·06	8·74	8·74	8·85	8·84	21·0	13	21·8	45
French Chalk	8·9	0·07	8·84	8·83	8·87	8·85	21·2	15	21·4	38
Alundum Cement	9·1	0·08	8·64	8·82	8·84	8·83	21·5	16	21·8	42

Tests have been reported[1] which showed that water-cooled copper-faced moulds were much superior to cast-iron for the production of strip bronze for rolling (Figs. 20, 21). These moulds ensure freedom from "blowing porosity," and due to the more effective chill, give finer and more uniform macrostructure and greater freedom

In some more recent work carried out since the completion of the investigations reported above, a resinous aluminium paint was developed as a mould dressing to prevent surface blowing in bronzes. This dressing was found to exert a more severe insulating effect between the mould face and the ingot surface, and reduced the rate of cooling to such an extent that the minimum pouring rate for a satisfactory 1-in. ingot of 2B8 bronze was reduced from 6 lb./min. at 1,200° C. (as determined in Section I of the present work, using tallow/charcoal dressing) to 4 lb./min. Details of some of these tests are given in Table IV.

(*c*) *Mould Material.*—The rate of cooling of an ingot during solidification, hence the extent of shrinkage porosity, will vary with the thermal characteristics of the mould material. In practice, however, cast-iron moulds are used almost universally for the production of solid or cored billets, and since the aim of these researches was to obtain control methods which could be applied directly to current industrial practice, no experiments were made on alternative mould materials for billet-moulds.

porosity (*i.e.*, porosity not related to shrinkage effects), the present tests were made with a low-phosphorus bronze which is less subject to this defect than the more commonly-used high-phosphorus bronzes.

Melts of the low-phosphorus 10% tin bronze were prepared as before, and poured at 8 lb./min., 1,250° C. (conditions for minimum shrinkage porosity as determined in preceding experiments) into the 1-in. mould, pre-heated to 100° C. The dressings tested included two refractory types

TABLE IV.

MODIFICATION OF CASTING CONDITIONS WHEN USING ALUMINIUM PAINT.

Mould Dressing	Pouring Rate	Casting Temp. °C.	Grms./c.c. Density as Cast		Grms./c.c. Density as Machined		Tensile Properties Top		Tensile Properties Bottom		Remarks
	Lb./min.		Top	Bottom	Top	Bottom	μ. T.S. Tons./sq. in.	Elon. % on 2 in.	μ. T.S. Tons/sq. in.	Elon. % on 2 in.	
Aluminium paint	5·3	1,150	8·71	8·64	8·71	8·64	28·0	16	25·3	11	Filled, good surface
Aluminium paint	5·1	1,175	8·79	8·73	8·79	8·74	28·0	20	25·2	12	Filled, good surface
Aluminium paint	5·7	1,150	8·82	8·79	8·77	8·62	25·7	11	24·1	11	Filled, good surface
Stockholm tar on chemically-cleaned mould	—	1,175	—	8·79	—	8·78	—	—	27·8	19	Struck badly, and did not fill
Aluminium paint	6·2	1,150	8·75	8·70	8·77	8·72	24·4	14	25·3	—	Filled, good surface
Stockholm tar on chemically-cleaned mould	5·7	1,175	8·79	8·77	8·80	8·78	28·4	27	28·0	24	Although filled, it gave slightly wrinkled surface
Stockholm tar on brushed mould	5·2	1,200	8·80	8·77	8·79	8·79	26·6	24	25·3	10	Filled, good surface
Stockholm tar on chemically-cleaned mould	4·9	1,200	8·79	8·77	8·77	8·73	28·4	20	25·0	10	Filled, good surface
Aluminium paint	6·6	1,175	8·74	8·70	8·76	8·69	24·1	11	25·1	18	Filled, good surface
Stockholm tar on chemically-cleaned mould	3·5	1,200	8·80	8·76	8·81	8·78	29·0	16	28·4	16	Although filled, it gave badly-wrinkled surface
Aluminium paint	3·9	1,175	8·78	8·77	8·79	8·80	28·2	23	26·5	18	Filled, good surface
Aluminium paint	3·6	1,150	8·78	8·77	8·80	8·80	27·6	17	26·6	13	Filled, good surface
Aluminium paint	3·9	1,100	8·80	8·75	8·81	8·76	27·8	17	27·4	16	Too cold, although it filled gave wrinkled surface
Aluminium paint	4·0	1,175	8·79	8·75	8·83	8·74	28·5	20	27·4	17	Filled, good surface

from porosity for equivalent pouring conditions. It is likely, therefore, that the critical minimum rate would be higher for a copper-mould than for a cast-iron mould. Similar advantages could be expected from the use of water-cooled or solid moulds of high-conductivity material, for the production of billets.

(*d*) *Effect of Mass of Mould.*—The effective mass of moulds for billets and sticks of which the height greatly exceeds the diameter, depends primarily on the thickness of the mould wall. The effect of variation of wall thickness was studied for billets of 2B8 bronze 12 in. long × 3 in. dia., from melts prepared as before and cast into moulds having 1 in., 1½ in. and 3 in. walls, respectively.

Preliminary tests with the 3 in. wall mould showed that the minimum pouring rate for complete filling, with pouring temperature 1,250° C., was 10 lb./min. Similar ingots were then prepared in all three moulds at this rate and at slightly slower rates down to 8 lb./min. These tests showed that the minimum rate did not vary with the mould wall thickness over the range examined—*i.e.*, 1 in.-3 in.

The ingots were sectioned and examined as shown in Fig. 2.

Microstructures show that variation of mould-wall thickness has no appreciable effect on the extent or distribution of porosity, but as mould thickness decreases the complex areas ($\alpha + \delta + Cu_3P$) decrease in extent, and tend to change in form from intercrystalline envelopes to isolated areas, and to become coarser in structure.

For equivalent pouring rate, density and tensile strength do not vary appreciably with mould-wall thickness (Table V), but elongation values increase as wall thickness decreases, in accordance with the observed improvement in compound distribution.

The relatively small effect of mould-wall thickness on porosity is in agreement with Bailey's findings on the effect of mould-wall thickness on the time of solidification of brass ingots[5] and can be explained by considering the low conductivity of the cast-iron mould. The rate of cooling during the relatively brief period in which solidification occurs (which determines the extent of shrinkage porosity) will be determined by the temperature gradient from the inner to the outer surface of the mould. It was shown by Bailey[5] that with C.I. moulds this gradient immediately after pouring is very steep for a short distance from the inner surface, and then flattens rapidly. Thus, any variation in thickness of the mould above a certain critical value (which is less than the thinnest wall tested in the present work) would have no effect on shrinkage porosity). A mould wall thinner than about 1 in., however, would be insufficiently rigid for practice, hence no tests were made with thinner walls. The observed variation of coarseness of the compound occurs during the period of cooling from the solidus towards room temperature. This time would be appreciably longer than the solidification period, and the mass of the mould becomes an appreciable factor, since the thinner the mould, the higher temperature it reaches, hence the slower the cooling of the bronze, with a resultant decrease in extent and coarsening in structure of the ($\alpha + \delta$) eutectoid.

Conclusions on the Effects of Mould Conditions

Variation of mould conditions such as temperature, wall thickness, and surface dressing have little effect, compared with that of pouring conditions, on the extent and distribution of shrinkage porosity, but have an appreciable effect on surface quality and structural condition of the bronze.

In general, mould temperature and thickness should be controlled with respect to ease of manipulation and the necessary rigidity, while mould dressings should be selected (as in the present practice) for control of surface quality.

Part III—The Relation Between Size and Shape of Ingot, and the Critical Pouring Rate for Maximum Soundness

In order to determine the general validity of the conclusion from the preceding work "that maximum soundness is obtained by pouring at the minimum rate which will fill the mould completely," a number of tests were made to determine the minimum pouring rate for ingots of varying size and form, and to examine the properties of such ingots.

Three main types of ingot were studied:

(*a*) Solid billets varying in diameter from 1–3 in.

(*b*) Cored billets varying in external diameter from 2–5 in., in internal diameter from 1–3½ in.

(*c*) Rectangular strips and bolts of dimensions: 36 in. × 12 in. × 1 in., 12 in. × 6 in. × 1 in., 12 in. × 4½ in. × ¾ in., 12 in. × 3 in. × ¾ in., 12 in. × 2 in. × 1 in., 36 in. × 2½ in. × 2½ in., and 12 in. × 1⅝ in. × 1⅝ in.

(*a*) *Solid Billets.*—Melts of 2B8 bronze were prepared as before, and were poured into the various moulds at 1,200°–1,250° C. The pouring rate for each billet (single central stream), was varied to determine the minimum rate which would fill the mould completely without cold shuts. All moulds were pre-heated to 100° C., and dressed lightly with tallow/charcoal.

The ingots were sectioned either as in Fig. 1, for ingots up to 2 in. dia., or as Fig. 2 for ingots 2½–3 in. dia. Microstructures were examined, and density, hardness and tensile properties were determined.

The minimum critical rate for each ingot bears a linear relation to the

5 G. L. Bailey. *J. Inst. Metals*, 1932, 49, 203.

TABLE V.

EFFECT OF MOULD WALL THICKNESS ON THE PROPERTIES OF 2B8 PHOSPHOR BRONZE.

Ingot Dia. Inches	Mould Wall Thickness. Inches	Pouring Aperture. Dia. Inches	Rate. Lb./min.	Densities. Grms./c.c.			Tensile Properties				D.P. Hardness	
							Centre		Edge			
				Whole Cast	Centre Mach.	Edge Mach.	U.T.S. Tons/sq. in.	El. % On 2 in.	U.T.S. Tons/sq. in.	El. % On 2 in.	Centre	Edge
3	2½	⅞	24	8·70	8·75	8·68	17·5	5	26·4	22	102	106
3	2½	15/16	16	8·72	8·79	8·62	20·9	6	26·7	21	120	112
3	2½	1	12	8·73	8·79	8·72	23·4	16	27·2	31	104	116
3	1	1	12	8·75	8·70	8·79	27·0	18	26·6	31	117	114
3	1	½	8	8·74	8·77	8·67	24·6	17	25·5	17	128	125

TABLE VI.

EFFECT OF INGOT SIZE ON THE PROPERTIES OF 2B8 PHOSPHOR BRONZE.

Ingot Diam. Inches	Pouring Aperture. Diam. Inches	Rate. Lb./min.	Densities. Grms./c.c.			Tensile Properties				D.P. Hardness	
						Centre		Edge		Centre	Edge
			Whole Cast	Centre Mach.	Edge Mach.	U.T.S. Tons/sq. in.	El. % On 2 in.	U.T.S. Tons/sq. in.	El. % On 2 in.		
3	2	8	8·75	8·70	8·79	27·0	17	24·5	35	115	114
2½	2	10	8·74	8·76	8·75	27·3	30	28·9	38	109	104
2½	1¼	6	8·77	8·78	8·77	29·8	42	28·3	34	148	132
2	1¼	6	8·80	8·78	8·82	27·2	26	26·7	21	119	110
2	1	8	8·74	8·76	8·68	27·5	31	26·6	22	110	109
1½	1	6	8·76	8·75	—	26·0	9	—	—	122	120
1½	¾	4	8·80	8·75	—	26·9	7	—	—	—	—
1	¾	4	8·78	8·76	—	28·0	15	—	—	120	118
1	½	4	8·78	8·76	—	27·5	15	—	—	—	—

diameter (or periphery) of the ingot (Fig. 22).

i.e., $R = KP$

where R = pouring rate.

P = periphery.

K = Constant.

The constant K, will involve the characteristics of the alloy and of the mould, and the degree of superheat—*i.e.*, the pouring temperature.

Examination of the billets made at various rates (See Table VI) confirms that density and properties are highest and most uniform across the section when the minimum pouring rate is used. Under these conditions there is no appreciable difference in density or strength as between an ingot 1 in. in diameter and one 3 in. in diameter. Elongation values, however, increase with increase in diameter of the ingot, and microstructures show that this is caused by a decrease in the amount of compounds present, due to the slower rate of cooling in the solid state. (Compare Figs. 28 and 29.)

With pouring rates above the critical value, central unsoundness occurs with attendant poor mechanical properties, while with rates below the critical the ingots show surface cold shuts and microfissuring near to the surface.

The formula relating the critical rate for maximum density to ingot size can be explained theoretically as follows. The condition for minimum shrinkage porosity in a given alloy is that the length of the pasty zone during solidification shall be a minimum. The extent of this zone is determined by the thermal gradients existing during solidification; the temperature gradient will be determined by the relation between the heat input and the heat extracted. But for constant degree of superheat (*i.e.*, constant pouring temperature) and constant material and temperature of the mould, the heat input in unit time is determined by the rate of pouring, and heat extraction by the area of mould surface in contact with the solidifying metal. The latter is proportional to the periphery of the

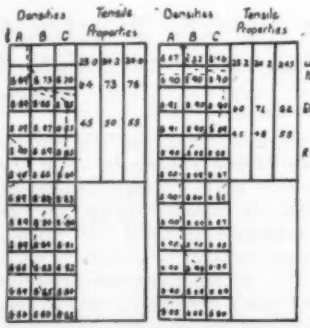

Fig. 20.—Variation of density and tensile properties in iron mould ingot.

Fig. 21.—Variation of density and tensile properties in copper-mould ingot.

Density contours at intervals of ·05 grm./c.c.

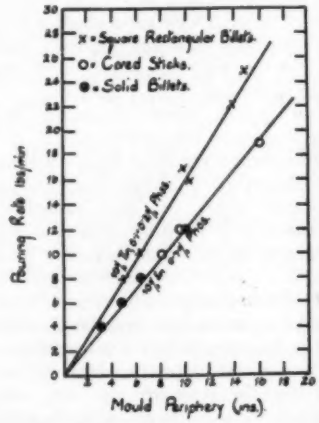

Fig. 22.—The relation between critical pouring rate and ingot size.

TABLE VII.

EFFECT OF POURING CONDITIONS ON PROPERTIES OF CORED BILLETS OF PHOSPHOR BRONZES.

Alloy		Billet Size. Inches.		Pouring Rate. Lb./min.	Pouring Streams	Pouring Temp. °C.	Remarks on Billet Surface	Density. Grms./c.c.		Tensile Properties				D.P. Hardness	
								Stream	Remote	Stream		Remote		Centre	Outside
Sn %	P %	Internal Diam.	External Diam.							U.T.S. Tons/sq.in.	El. % on 2 in.	U.T.S. Tons/sq.in.	El. % on 2 in.		
9·8	0·70	1½	2	15*	2	1,250	Good	8·74	8·70	26·6	14	24·4	10	112	107
10·4	0·77	1½	2	16	2	1,250	Good	8·62	8·74	26·0	12	22·8	8	121	123
10·5	0·75	2½	3	24*	2	1,250	Good	—	8·42	24·5	8	22·5	6	—	—
10·5	0·75	2	3	36	3	1,250	Good	—	—	21·9	10	18·6	6	—	—
10·5	0·75	2	3	16	2	1,250	Rumpled slightly	8·54	8·60	19·2	4	22·4	10	—	—
10	2% Zn	2	3	25*	2	1,200	Good	8·69	8·76	23·4	10	19·2	4	—	—
5 Sn	5 Zn 5 Pb	2	3	25*	2	1,200	Good	8·67	8·68	16·8	6	16·8	6	—	—
15	0·1	3½	5	22	2	1,200	Good	8·79	—	20·0	2	17·4	1	—	—
10·3	0·8	3½	5	35*	4	1,200	Good	8·72	8·68	26·0	18	23·3	13	—	—
9	1·5	3½	5	24*	4	1,200	Good	8·53	8·50	28·0	17	25·6	11	—	—
9	1·5	3½	5	48	2	1,200	Rough	8·60	—	26·0	11	24·0	7	—	—
9	1·5	3½	5	24	2	1,200	Good	8·60	8·62	28—30	10—30	24—29	10—17	—	—

* Critical pouring rate.

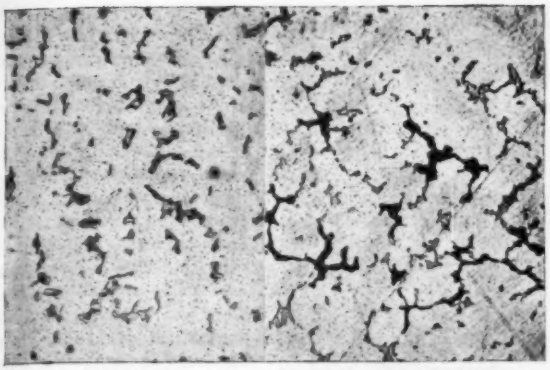

Fig. 23.—10% tin bronze. Fine pinhole shrinkage cavities in isolated ($\alpha+\delta$) areas. Poured at 1,250° C., minimum pouring rate. × 150.

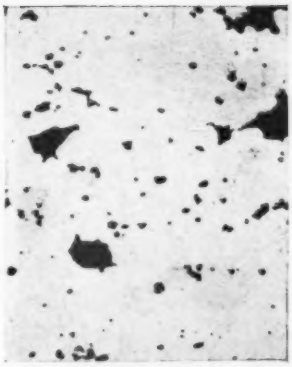

Fig. 24.—10% tin bronze. Fissure-porosity. Rapid pour and sluggish metal. × 150.

Fig. 25.—5% tin bronze. Coarse shrinkage cavities. Poured very rapidly. × 150.

ones, it was found preferable to use a greater number of streams—*e.g.*, four for billets 2–4 in. external diameter, six for billets 4–6 in. external diameter.

Using multiple-stream pouring, it was found that the minimum pouring rate to avoid cold shuts in a cored billet was much greater than that for a solid billet cast in the same mould. This is to be expected, since the cored billet solidifies in contact with a much larger chilling surface, viz., the sum of the surface areas of the

ingot. Then, for a minimum constant pasty zone:

Heat input = Heat extracted.

but heat input = K_1R

and heat extracted = K_2P

where R = cooling rate,

P = periphery of ingot.

$K_1R = K_2P$

$R = K_3P$

(*b*) *Cored Billets.*—Melts of 2B8 bronze were prepared as before, and were poured at 1,200° C. into the various moulds. In each test the core was of cast iron coated with a layer of loam $\frac{1}{32}$ in. thick, consisting of 5% Kordol (an oil-cereal binder) with sea sand. This was baked at 200° C. before use, since this procedure was known from previous investigations[6] to minimise gas absorption from the loam.

In determining the optimum pouring conditions with cored billets, one extra variable must be taken into account, viz., the number and positioning of the pouring streams employed. Preliminary tests showed that a single stream tended to cause cold shuts at the side of the casting opposite to the position of the stream, unless a very rapid rate of pouring was used, which in turn caused shrinkage porosity at the position of the stream. Using two pouring streams diametrically opposite eliminated this trouble with small-cored billets (up to 2 in. outside diameter), but, with larger mould and of the core. The variation of the critical rate with ingot size

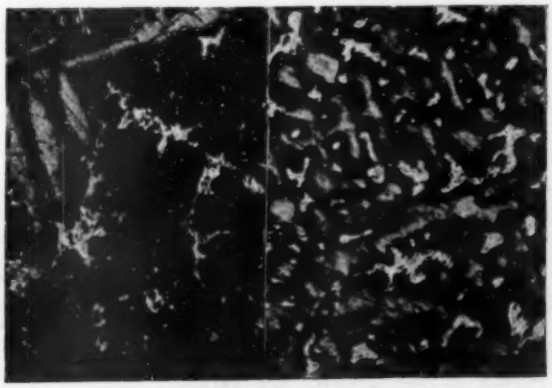

Fig. 26.—2B8 bronze (10% tin, 0.75 phos.). Poured into cold mould (20°C.). Intercrystalline ($\alpha+\delta$). × 150.

Fig. 27.—2B8 bronze (10% tin, 0.75% phosphorus). Poured into hot mould (200° C.). Isolated areas of ($\alpha+\delta$). × 150.

6 W. T. Pell-Walpole. *Foundry Trades Journal*, 1946, 78, 87.

TABLE VIII.

RELATION BETWEEN COMPOSITION, MINIMUM POURING RATE AND MECHANICAL PROPERTIES OF COMMERCIAL TIN BRONZES.

Analysis.					Pouring Conditions.			Density gm./cc.		Tensile Properties				D.P. Hardness		Value of K in Formula R = KP
										Top		Bottom				
Sn	Zn	Pb	Phos.	Ni	Temp. °C.	Rate Lb./min.	Aperture Diam.	Top	Bottom	U.T.S. Tons/sq.in.	El. % on 2in.	U.T.S. Tons/sq.in.	El. % on 2 in.	Centre	Edge	
10·6	—	—	0·73	—	1,250	8+	¼	8·78	8·76	27·5	31	26·6	22	119	110	1·3
10·5	—	—	2·0	—	1,200	8	¼	8·67	8·46	29·3	15	25·7	4	—	—	1·1
10·0	—	—	1·9	—	1,200	7+	5/16	8·67	8·53	30·6	15	29·6	10	123	130	1·1
15·0	—	—	0·65	—	1,150	7+	5/16	8·69	8·72	28·3	5	27·2	3	160	180	1·1
10·2	1·8	—	0·04	—	1,200	16	5/16	8·78	8·79	18·6	10	16·6	8	96	96	—
10·3	2·1	—	0·05	—	1,200	+12	5/16	8·86	8·86	22·0	18	22·6	16	93	92	1·9
10·0	2·0	—	0·04	—	1,200	+12	5/16	8·82	8·82	19·8	11	20·8	13	89	93	—
10·0	2·0	—	0·07	—	1,150	12	5/16	8·82	8·80	19·9	14	18·6	11	—	—	—
10·0	2·0	—	0·10	—	1,150	10+	5/16	8·83	8·82	21·7	16	20·1	12	86	104	1·6
5·0	5·0	5·3	0·02	—	1,200	8+	¼	—	8·67	—	—	18·6	36	—	—	—
5·6	4·5	5·6	0·01	—	1,200	12+	5/16	8·63	8·81	16·4	28	15·7	8	66	72	1·9
5·2	4·7	4·9	0·08	—	1,200	10+	5/16	8·88	8·87	19·0	33	19·0	32	86	91	1·6
5·3	4·6	5·1	0·19	—	1,200	8+	¼	8·64	8·86	20·4	38	19·7	32	76	76	1·3
7·6	2·4	2·6	0·63	0·9	1,200	8+	¼	8·84	8·89	25·3	28	23·9	16	107	112	1·3
10·2	—	9·3	0·07	—	1,200	10	5/16	8·83	8·84	17·6	7	17·4	7	80	72	—
9·7	—	8·4	0·05	—	1,200	8+	¼	8·94	8·94	19·4	16	16·4	13	82	87	1·3
4·6	—	19·0	0·06	1·0	1,250	8	¼	9·10	9·10	—	—	15·4	17	—	—	—
5·6	—	18·5	0·06	1·0	1,250	12+	5/16	9·10	9·10	15·8	22	15·2	19	63	63	1·9

+ critical pouring rate.

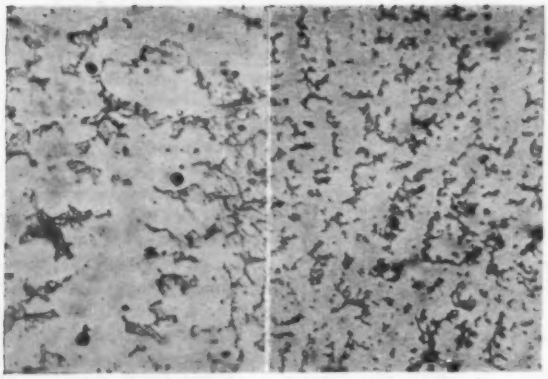

Fig. 28.—2B8 bronze (10% tin, 0.75% phosphorus). Poured at minimum rate into 3 in. dia. mould, 1 in. walls, at 1,100° C. Coarse (α+δ), in isolated areas. Some air traps. × 150.

Fig. 29.—2B8 bronze (10% tin, 0.75% phosphorus). Poured at minimum rate, 1,150° C., into 1½ in. dia. mould, 2 in. walls. Fine intercrystalline (α + δ). Trace of blowing porosity. × 150.

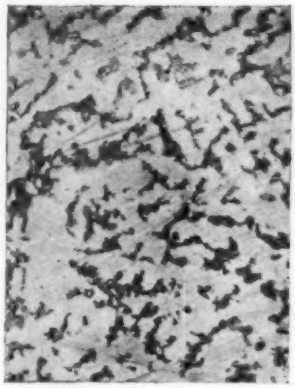

Fig. 30.—High phosphorus bronze (10% tin, 2% phosphorus). Poured at critical rate into 2 in. dia. mould with 1¼ in. dia. core. No porosity. × 250.

(Fig. 22) is such that

$$R = K\ (P_i + P_e)$$

where R = Pouring rate,
P_i, P_e = Internal and external peripheries respectively of the casting.
K = A constant.

Examination of the cored billets (Table VII) confirmed that those poured at the critical rate give optimum density and mechanical properties, though these are slightly inferior to the corresponding values for solid sticks poured in the same moulds. This is probably due to slight gas porosity caused by absorption of hydrogen from the loam core-coat.[8] Considerable variation in mechanical properties occurs throughout cored billets cast with an insufficient number of streams. With slow rates of pouring, the properties at the positions of the pouring streams are superior to those at positions remote from the pouring streams, while with rapid rates the reverse is true. An appreciable annealing effect occurs around the position of the stream, producing better properties, provided that local porosity is absent—*i.e.*, when the pouring rate is slow. Rapid rates cause sufficient porosity to counteract the improvement produced by the annealing effect.

(*c*) *Ingots of Square or Rectangular Section.*—Since ingots of square or rectangular section are generally made for rolling, tests with these shapes were made with alloys suitable for fabrication—*i.e.*, having a lower phosphor content than the 2B8 casting bronze used in the preceding tests. Melts of phosphor bronze containing 7–12% tin with 0·1–0·2% of phosphorus were prepared under a degassing flux, and were poured at 1,200° C. (into the moulds listed previously). The square bolts and strips up to 6 in. wide were poured with a single central stream, since it has been shown previously that this gives better results than multiple stream pouring, but with ingots 12 in. wide, a double stream was found preferable, since this permitted the use of a lower total pouring rate, while avoiding the occurrence of cold shuts on the ingot edges.

The relation between ingot size and minimum critical pouring rate (shown in Fig. 22) upper curve, again has the form

$$R = KP$$

where R = pouring rate,
P = ingot periphery,
K = a constant.

The slope of the line—*i.e.*, the value of K, is different from that for the high-phosphorus bronze, as might be expected from the different characteristics of these alloys.

These ingots were subjected to cold-rolling tests. Those poured at the minimum rates all rolled successfully without cracking, while traces of edge cracking and occasional axial cracks

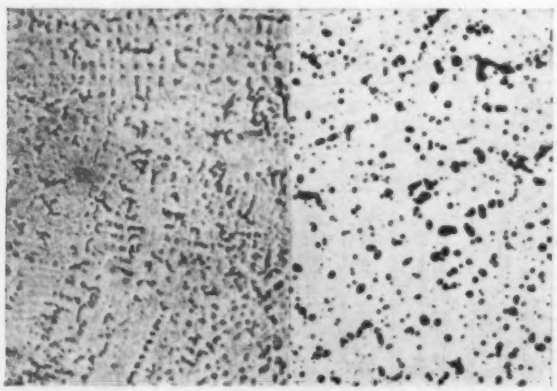

Fig. 31.—Gun metal (88/10/2). Cast at critical rate. Sound. × 150.

Fig. 32.—Leaded gun metal (85/5/5/5). Cast at critical rate. Sound. Fine uniform lead distribution. × 150

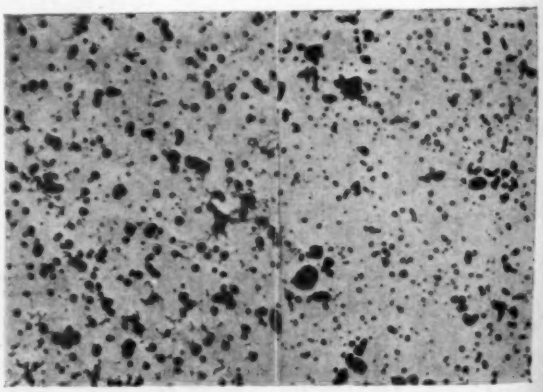

Fig. 33.—Leaded bronze (10% lead, 10% tin). Poured at critical rate. Sound. Uniform lead distribution. × 150.

Fig. 34.—High-lead bronze (20% lead, 15% tin). Poured at critical rate. Sound. Uniform lead distribution. × 150.

occurred in ingots poured more rapidly, especially in those of high tin content.

Conclusions on Effect of Size and Shape of Ingot

Maximum soundness, mechanical and working properties are obtained in any ingot irrespective of size or shape, provided it is poured at the minimum rate which will fill the mould without cold shuts. This critical rate varies with ingot size according to the formula

$$R = KP$$

where R = Pouring rate,
P = Periphery of ingot cross-section, or in the case of cored billets $P = (P_i + P_o)$—*i.e.*, the sum of the internal and external peripheries of the casting.
K = Constant which varies with the characteristics of the alloy.

Part IV—The Relation Between Alloy Composition and the Critical Pouring Rate for Maximum Soundness

In the detailed investigations of the effect of pouring conditions on the density and mechanical properties of various bronzes (Parts I and III), it was noted that the minimum pouring rate for optimum properties was lower for the more fluid high-phosphorus bronze than for the low-phosphorus bronze, gun metals and leaded bronzes —*i.e.*, the value of K in the formula R = KP varies with the alloy composition.

In order to enable this formula to be applied to the full range of tin bronzes as normally used, a series of tests was made to determine the minimum pouring rate, hence the value of K for ingots 12 in. × 2 in. dia. in a series of alloys. The results and properties of the ingots prepared are given in Table VIII.

The phosphorus content of a bronze has the greatest effect on the value of the constant K, which decreases from 1·6–0·9, as phosphorus content is increased from 0·05%–2%. The value of K also decreases slightly with increase of lead or of tin content, and increases with increase in zinc content; thus, for constant mould conditions, the critical pouring rate decreases as the fluidity of the alloy increases.

The properties of the ingots of the various alloys examined (Table VIII) confirm that for each composition the maximum density, highest and most uniform properties across the section are obtained with the minimum pouring rate. Microstructures of alloys poured under these conditions are shown in Figs. 23, 29–34.

Part V—General Conclusions

Maximum soundness and highest and most uniform properties throughout the ingot section are obtained in degassed, chill-cast bronzes, gun metals and leaded bronzes, provided the metal is poured at the minimum rate which will fill the mould without cold shuts. This critical rate is related to ingot size by the formula

$$R = KP$$

where R = pouring rate.
P = periphery of cross-section of a solid billet, strip or bolt, or in the case of cored billets is the sum of the external and internal peripheries of the cored billet.
K = A constant which varies with the alloy composition, being lowest for the fluid high-phosphorus bronzes and highest for low-tin high-zinc and high-lead bronzes.

The formula assumes that pouring temperature, mould temperature, and mould material and dressing are constant. Variation of these factors within practicable limits has little effect on soundness, as compared with the effect of pouring rate, though they affect surface quality and mechanical properties, the latter by causing variation in the amount and mode of distribution of the compound phases. The most convenient pouring temperature is 1,200° C., mould temperature = ~ 100° C. The most suitable dressing for cast-iron moulds is a resinous aluminium paint containing ~ 20% aluminium.

Acknowledgments

This work was carried out at Birmingham University, in collaboration with the Tin Research Institute. Special acknowledgment is due to two members of the Institute's staff, Dr. K. Winterton, for help with some of the experimental work, and Dr. J. W. Price, who carried out the chemical analyses.

Approval for publication has been granted by the Director-General of Research (Defence).

Ductile Zirconium from Zircon Sand

By W. J. Kroll, A. W. Schlechten and L. A. Yerkes

A PROCESS is described for making ductile zirconium metal using zircon sand as a starting material, the essential steps of which are: (1) Production of carbides from zircon sand; (2) chlorination of the carbides; (3) sublimation of the chlorides; (4) reduction of the chloride with magnesium; (5) vacuum distillation of the reaction products; (6) melting the metal in a vacuum arc furnace. The zirconium ingots obtained have a bright smooth surface and a Brinell hardness as low as 197. All ingots with a hardness less than 300 Brinell can be rolled into sheet. They are heated to 700° C. and rolled hot for the first few passes; after the thickness has been reduced somewhat, the strip can be cold rolled with annealing after several passes. The resulting sheet can be bent cold 180° without cracking. The solid zirconium can be heated to 700° C. in air without burning, since a protective oxide coating forms over the surface.

The process is not limited to the use of magnesium as a reducing agent; calcium, lithium and sodium could also be used. These other metals have certain disadvantages; calcium and lithium tend to be impure, as they oxidise readily and their nitrides are soluble in the metal. After the reduction operation it would be difficult to remove the $CaCl_2$, as its vaporisation point is above 1,200° C. The use of sodium creates a considerable fire hazard, and there is a possibility that it would be lost in the reduction process if the temperature went above its boiling point of 880° C.

Zirconium is not the only metal capable of being produced by this method; the procedure and equipment can be used for the reduction of $TaCl_5$, $CbCl_5$, $BeCl_2$, and UCl_4, which have boiling points of 233°, 241°, 488° and 618° C., respectively, and for some others depending upon the volatility of their salts. Separation of the chloride from the oxide is not as important for some of these metals as it is for zirconium and titanium if they are not embrittled by the presence of oxide, but it would be especially helpful in the case of beryllium, where the presence of oxide makes remelting most difficult.

From *Electrochem. Soc.*, Preprint, 89–29, April, 1946.

Petroleum Distillation Extracts as Core-Binders

IN view of the increasing shortage of linseed oil, considerable attention has been given to the possiblity of replacing its use in core-binders by other oils in greater supply. The use of aromatic extracts removed by solvents from the lubricating oil fraction of crude petroleum has been given special consideration by Dr. Davies and Dr. Rees, at Sheffield University, in collaboration with the Petroleum Board Technical Staff. The results of the investigation, published in a recent issue of the Journal of the Iron and Steel Institute, indicate that these extracts can be used to reduce the linseed oil generally used in core sands and the fact that they are available in considerable quantities and are of uniform composition facilitates their use. The following conclusions from the above-mentioned investigation indicates more clearly the results obtained.

In sand mixes bonded with 2% linseed oil up to 40% of the linseed oil can be replaced with any of the extracts examined, except one, without any falling off in tensile strength. The exception, that designated Group 1 A (140), causes a gradual drop in tensile strength from 392 lb./sq. in., the value for linseed alone, to 335 lb./sq. in. at 40% replacement. With all the extracts, a rapid drop in tensile strength occurs when over 50% of the linseed oil is replaced by extract, and with over 50% replacement surface friability after drying develops.

With up to 40% of extract in the oil, however, both strength and surface firmness are equal to those with linseed oil alone. The optimum drying treatment for A.F.A. test pieces bonded with the mixtures was found to be two hours at 200° C. The milling conditions required are similar to those for mixes bonded with linseed oil alone, except in the case of a few of the most viscous extracts which need rather more milling. Extract Group 1 (72), however, which gave the best all-round properties, is of a consistency that makes it easy to handle. Premixing of the extract and linseed oil are not necessary, though doing so lessens slightly the time required to obtain a uniform mix.

For core compounds rich in starch up to 50% of the linseed oil can be replaced by extract, giving increased green compressive strength without materially altering dry compressive or tensile strengths. No surface friability results.

For core compounds rich in dextrin up to 40% of the linseed oil can be replaced by extract, giving increased green compressive strength, but a slightly lower dry compressive and tensile strengths. No surface friability results.

For core compounds rich in molasses replacement of 50% of the linseed oil gives an increase in green compressive strength, but lowers dry compressive and tensile strengths. After drying, the surfaces show slight friability.

For core compounds rich in sulphite lye replacement of 50% of the linseed oil reduces the dry compressive and tensile strengths.

The replacement of linseed oil by an extract/linseed oil mixture containing 40-45% extract does not alter the stripping qualities of sand mixes.

Trials carried out in two steel foundries confirmed those findings. In one of those foundries 40% of a commercial core oil was replaced by extract Group 2 (2239) with complete success. At another 40% of the linseed oil in a commercial core compound was replaced by extract Group 1 (72) with completely satisfactory results. At another foundry where cores are normally skin-dried at 140° C. for one hour, an increase in the number of rejects was reported following partial replacement of linseed oil by extract in the core mixture. Tests show that complete drying out is necessary to get optimum results with extract/linseed oil mixes, two hours at 200° C. being sufficient for A.F.A. test-pieces.

Mr. W. B. Parkes of the British Cast Iron Research Association, has drawn attention to some difficulties encountered in certain iron foundries due to distortion in the green state of core assemblies when petroleum distillation extracts were constituents of the core oils. It is possible that these difficulties may be eliminated by the use of high-viscosity petroleum extracts admixed with linseed oil, and it is understood that trials in this direction are in progress.

Spitfire Mitchell Memorial Scholarships

SEPTEMBER 15th is the sixth anniversary of the climax of the Battle of Britain, when 185 enemy machines were shot down. It was the Battle which, in the opinion of many of the best observers, saved the civilisation of the world. Much of the credit for this victory was due to that remarkable machine, the Spitfire, which was originally designed by R. J. Mitchell. As a memorial to his great work, two Scholarships have been founded to train students in aircraft design and engineering. These Scholarships will be of the value of £60 per annum and will be awarded for a period of three years at University College, Southampton. It is intended to make the first of these awards in time for successful candidates to begin their training at Southampton in October this year.

Applications for these Scholarships should be made to the Registrar, University College, Southampton, not later than August 12th, 1946, and should be accompanied by a statement of the candidate's education and experience, if any, and a recommendation from the candidate's employer or headmaster, together with the candidate's own age, which must be over 17. Candidates will be expected to sit for an examination in mathematics and an essay on subjects selected by them from a wide choice, and it is expected that those candidates who satisfy the examiners will be interviewed some time early in September. It is intended by these Scholarships to choose candidates who will, it is hoped, follow closely in Mitchells' footsteps. It was at Southampton that he did so much of his remarkable work and it is to Southampton that these young men will go.

The Royal Aeronautical Society, the Society of British Aircraft Constructors, Messrs. Vickers-Armstrongs, Ltd., and the Committee of the Spitfire Mitchell Memorial Fund will be responsible for choosing these young men.

HEAD, WRIGHTSON & Co. Ltd., Thornaby-on-Tees, announce that they now have an agreement with the Arthur G. McKee Company, of Cleveland, Ohio, U.S.A., by which they can call upon the McKee Company for engineering services in the design and construction of iron and steelworks plant in Great Britain and South Africa. Head, Wrightson also have the sole agency for supplying "Brosius" specialities in this country.

ONE feels inclined to wonder whither the recent developments in science in this country are leading. By this is meant, not the actual scientific achievements, but what one might refer to as the "organisational" side of things. We were pleased when, some time ago, the Microchemistry Group of the Society of Public Analysts was formed. This was followed rapidly by the formation of a Physical Methods Group and a Microbiological Group. Now, in the present issue, we have pleasure in reporting on the activities of the Industrial Spectroscopy Group of the Institute of Physics, which, it may be remarked, produced much of value for the chemist. But we are somewhat disturbed, nevertheless, lest in this multiplicity of Groups liaison may lapse. Thus, one member at least of the Physical Methods Group, interested in spectroscopy, was disturbed to learn that the Microchemistry Group also counts spectroscopy to some extent within its purview, since it deals with the analysis of small amounts of material. Recently, we learn, the Physical Methods Group held a symposium on spectroscopy in Edinburgh. Now we have this Conference in London. All these are undoubtedly encouraging signs of the getting together of scientific men. But we must be careful lest, in meetings as well as in literature, we may shortly need public pleas for some central clearing house, which will avoid duplication of material, and which will display less wealth to more advantage.

The Application in Routine Chemical Laboratories of the Alcoholic Iodine Method for the Separation and Analysis of the Oxides in Steel

By B. Rogers, A.I.M.

Residues obtained from rimming steels by alcoholic iodine extraction methods normally weigh 6–10 mg. Very reliable procedures for the analyses of the residue must be used to give satisfaction. A scheme of analysis is here proposed. Notes are appended on the scheme and on experimental alternative procedures suggested for obtaining information on the mineralogical formation of the oxides present in the steel.

THE method of extraction devised by the United Steel Co.[1] and known as the simplified boiling method is used to extract the oxides from the steel. The selected analytical scheme enables the whole of the solution to be used for the colorimetric determination of titanium and the volumetric determination of iron in sequence; manganese is determined by a Spekker colorimetric method; and aluminium by the aluminon procedure.

The use of colorimetric methods is becoming popular. The thioglycollic acid procedure[2] is the most useful one tried for the determination of iron. At the moment, gravimetric methods of determining iron with cupferron are not regarded as being very suitable by the writer, for this class of work. The determination of aluminium is best made colorimetrically by the procedure using aluminon[3].

Titanium is present in such small amounts that gravimetric methods are of little value owing to coprecipitation of titanium with iron, using sulphuretted hydrogen, and to the presence of impurities in the reagents. There is no point in determining manganese by a series of determinations on precipitates and filtrates when a good colorimetric method is available.

Chromium is present in very small amounts; and a colorimetric method applied directly to one acid solution is available. As graphs have to be prepared, the chemist in routine laboratories might adopt the following simpler routine scheme.

The Apparatus for the Separation of the Oxides

(*a*) The apparatus used for the extraction of the residue is a duplicate of the apparatus with ground

1 *J.I.& St. Inst.*, 1941, Part B, p. 350. Stevenson and Speight. Simplification of the Alcoholic Iodine Method for the determination of the oxides in steel.
2 I.& St. Inst. Spec. Report No. 25, p. 179. W. W. Stevenson. The analysis of Non-Metallic Residues obtained by the Alcoholic Iodine Method.
3 *J. I.& St. Inst.* No. 1, 1941, p. 371. G. E. Speight. The analysis of Non-Metallic Inclusions by the Alcoholic Iodine Method.

glass joints described by Stevenson and Speight in the Third Report of the Oxygen Sub-Committee[1].

(*b*) For the purification of the alcohol, an apparatus with ground-glass joints was used which was already available. A.R. alcohol is redistilled once or twice and tested by the special hydrometer made for the purpose or by specific gravity determinations, correcting for errors due to temperature in the final calculations[4].

(*c*) The apparatus for purifying the nitrogen is largely as in the published methods[5,6]. Two nichrome wound furnaces are used with electric mains supply and suitable resistances or transformers to enable the desired temperatures of 650° C. and 850° C. to be maintained in separate silica combustion tubes (*a*) and (*b*). In sequence, pyrogallol solution, (*a*), copper gauze at 650° C., a sulphuric acid tower, (*b*), iron powder (a simple mercury safety gauge), and two anhydrone towers are used to purify the nitrogen and/or indicate the rate of flow of the gas.

(*d*) For filtration of the alcoholic extract of the steel, an ordinary universal glass filter pump (with mains pressure) is used with a filter flask with side arm, in the neck of which is placed a plain funnel with a filter disc or cone, covered by pulp made from one 44 Whatmans paper (15 cms.).

An Outline of the Selected Analytical Scheme

The residue is ignited in a platinum crucible and weighed on a semi micro-balance (after 25 minutes in a dessicator), and fused with potassium bisulphate. The sulphates are extracted with dilute sulphuric acid and silica is filtered off, ignited, weighed and volatilised as usual with hydrofluoric acid in the presence of sulphuric acid.

The residue is fused with potassium bisulphate and extracted as above.

Cupferron is used to separate iron and titanium from the other elements present and the filtered and washed precipitate thus obtained is ignited and fused with potassum bisulphate.

Titanium is determined by visual colorimetry on the whole solution with hydrogen peroxide. The solution is then transferred to a beaker for the volumetric determination of iron by reduction with stannous chloride and oxidation with potassum dichromate using diphenylamine indicator[7].

The filtrate of the cupferron precipitation is treated with an exact amount of sulphuric acid and some nitric acid in order to destroy cupferron and leave a solution suitable for a main solution of known volume. Fractions of this solution are taken for the remaining determinations. The determination of manganese is made by a Spekker colorimetric method, upon a quarter fraction of the solution[8]. On a one-fifth fraction, manganese is removed by ammonia precipitation of aluminium hydroxide[9]. The precipitate is washed and dissolved in dilute hydrochloric acid, and aluminium is determined by the colorimetric procedure using aluminium[3] on one-half of the acid solution.

Chromium is determined on the one-tenth fraction by fuming with sulphuric acid, peroxidising as for manganese[8,10], and adding diphenylcarbazide solution[2].

Phosphorus is determined by the alkalimetric phosphomolybdate procedure upon the remainder of the main solution.

Notes on the Procedures Used in the Selected Scheme of Analysis

(*a*) Iron and Titanium.

Iron is determined as set out in Lundell's text book on the chemical analysis of Steel (Section on Iron Ores)[7]. It is necessary to use N/100 dichromate solution and a 10 mls. burette graduated in $\frac{1}{50}$ths. The usual addition of phosphoric acid for the titration is omitted.

The yellow pertitanate disappears on boiling with stannous chloride. The first procedure A(1)[8] on test gave the following results indicating that a quantitative ratio is present, if not a 100% colour development. Tabulated results were obtained as follows on synthetic solutions :—

For Iron :—

	Expt. 1	Expt. 2	Expt. 3	Expt. 4
Wgt. of Fe_2O_3 taken in mg. ..	4·0	8·00	2·5	4·0
Wgt. of Fe_2O_3 obtained	3·92	7·32	2·35	4·08

This is a suitable range for good and poor quality wires of rimming steel type. Using a fresh indicator solution, this method was more satisfactory than gravimetric methods *or* colorimetric (thiocyanate and thioglycollic) methods, in the writer's experience.

For Titanium :—

Wgt. of TiO_2 taken in mg.	0·46	0·575	0·92
Wgt. of TiO_2 found in mg.	0·30	0·38	0·61

A 66% colour development is obtained by adding a standard solution of titanous sulphate to the blank solution and matching the pertitanate against suitable synthetic solutions. Sample solutions are compared therefore, against synthetic standard solutions.

(*b*) Aluminium.

1. From the fraction taken for the determination of aluminium, manganese is best removed in a volume of 35 ml. of a dilute hydrochloric acid solution containing a little ammonium chloride. The solution is reduced with a few drops of sulphurous acid, oxidised with only two drops of nitric acid, and boiled[11]. Dilute ammonia is added with methyl red as an indicator to obtain a very slight excess of ammonia, which may be boiled out in 1–2 mins. This period is considered to be better than long periods of digestion, particularly in the presence of large amounts of manganese[8,12].

2. Ammonia was omitted from the ammonium carbonate solution without ill effects being observed and there is no point in intensifying fading of the colour once formed.

3. It is essential to use the same stock solution of *water*, acids and reagent solutions for the blank, sample and synthetic standard solutions. Aluminon is superior to quinalizarin or to alizarin S as a reagent. Correction factors for iron are not applicable as if iron is not removed the results are irregular. In perchloric acid solution, the colour darkens steadily with the increased length of time of standing.

4 Fourth Report (Oxygen Sub-Committee, Section II., Part A (b), p. 20.) The effect of water on Methanol.

5 *J. I. & St. Inst.* 1935. Vol. 1. Spec. Report No. 9, p. 56. Purification of Nitrogen.

6 *J. I. & St. Inst.* 1937, p. 249. Rooney and Stapleton.

7 G. E. F. Lundell, J. I. Hoffman, H. A. Bright. Chemical Analysis of Iron and Steel. 1931. p. 528.

8 Selected Methods for the analysis of Rimming Steels and Weld Metals. B. Rogers. Unpublished.

9 See Notes on Procedures (*b*).

10 E. J. Vaughan. The use of the Spekker. Photo-electric Absorptiometer in Metallurgical Analysis. Monograph published by the Inst. of Chemistry, 1941. First lecture.

11 G. E. F. Lundell, J. I. Hoffman and H. A. Bright. Chemical Analysis of Iron and Steel, 1931, pp. 349 and 423.

12 Loc. cit., p. 68 in text.

(*c*) MANGANESE.

Manganese might be determined visually in Nessler tubes in the absence of a photometer.

(*d*) PHOSPHORUS.

Phosphorus is best collected by adding H.S. spectroscopic iron solution in excess and the orthophosphate is precipitated with ammonia. A solution nitric acid enables the application of the usual phospho-molybdate method. The alkalimetric procedure is preferred using N/50 sodium hydroxide and N/50 sulphuric acid with phenolpthalein as an indicator.

Suggestions for Improving Published Procedures of Analysis of the Residues

The Procedure I. (Chemists' Panel)[13] may be modified (especially for alloy steel residues) by boiling the acid solution used for the determination of aluminium with potassium persulphate, precipitating with 8 hydroxyquinoline in slightly acid solution and making the solution alkaline with ammonia.

This is claimed by Lundell to separate phosphorus and chromium etc. quantitatively[14]. If true for the inclusion analyses, this would enable a simplified analytical scheme to be adopted. For alloys, a portion of the persulphate peroxidised solution could be used for the determination of chromium using diphenylcarbazide, etc. and the remainder *used for aluminium and phosphorus determinations.*

A Discussion of Suggested Direct Methods of Analysis of Residues

(*a*) FOR INFORMATION CONCERNING THE MODE OF OCCURRENCE OF THE OXIDES IN THE STEELS.

It is thought that if suitably large residues (say 100 mg.) could be obtained by 10% sulphuric acid or 20% hydrochloric acid extraction methods, information of value concerning the form of occurrence of the oxides might be obtained by examination of the residues following the use of suitable solvents and liquids for selective gravity separation (*i.e.*, bromoform, etc.) in *sequence* as far as this is practicable, with examination of the residues at each stage by physical and chemical tests.

(*b*) A BASIS FOR A DIRECT METHOD OF ANALYSIS.

1. Gravity separations—bromoform, cadmium boro tungstate, mercuric potassium, iodide, etc. (See Schoeller.)[15]

2. (*a*) 50% acetic acid solution is a solvent for phosphates and metallic manganese and iron; and, partially for oxides of manganese; and has been used for flux analyses. (*b*) Nitric acid, with hydrogen peroxide, is also similarly capable of dissolving oxides of manganese and residues generally[16, 17].

3. Magnetic tests may confirm the presence of spinels[15]. Dipyridil[18] might be useful to determine ferrous oxide, quantitatively at this and other stages of these group separations.

4. Nitric acid with or without hydrofluoric acid, in which alumina is insoluble[19], may lead to direct methods by enabling alumina to be removed. (See (*c*) below.)

13 Loc. cit., p. 351 in text.
14 I. & St Inst. Spec. Report No. 25, p. 184. W. W. Stevenson. The analysis of Non-Metallic Residues obtained by the Alcoholic Iodine Method.
15 W. R. Schoeller and A. R. Powell. The Analysis of Minerals and Ores of the Rarer Elements. Ch. II., p. 8.
16 In routine use in Murex Laboratory. B. Rogers.
17 W. Scott. Standard Methods of Chemical Analysis. Fifth Edition. Vol. 1, p. 564.
18 R. Hill. Proceedings of the Royal Society. 107B, 205, 193–.
19 J. W. Mellor. Inorganic Chemistry. Vol. V., Al, p. 270.

5. Alcoholic solutions of hydrochloric acid attack ferrous oxides and probably manganous oxides[3, 20].

(*c*) A DIRECT METHOD IN OUTLINE FOR THE DETERMINATION OF ALUMINA AS SUCH.

These latter tests, b4 and b5, will, if quantitative, suggest another possibility, viz., that it might be possible to remove all oxides except alumina itself and get a direct gravimetric determination of aluminium. As the 20% hydrochloric acid extraction procedure[21] is stated to yield 95% of the alumina, it should be practicable following an alcoholic iodine extraction, to obtain satisfactory results provided solubility tests of alumina support this claim[19]. Alumina formed at high temperatures is the most difficult alumina to dissolve in acids[19] and therefore open hearth and weld metal (particularly acid slag inclusions and normal steel[21]) should be analysed successfully by this procedure for aluminium. The filtrate having been passed through a suitable filter (porous crucible) could be used for the direct colorimetric determinations of the other elements.

The Use of a Direct Confirmatory Etching Test on the Steel Sample

Metallographic tests are of course already used, whilst it is interesting to note that L. Fine[22] believes that ferrous oxide may be shown up by an alkaline chromate etch without analyses by Inclusion Residue Methods and this may offer promising results.

Acknowledgments are made to the Board of Directors of Murex Welding Process, Ltd., for permission to publish this paper, and to Dr. E. C. Rollason, Research Director, for encouragement.

20 Metals Handbook.
21 J. F. Lundell, J. L. Hoffman and H. A. Bright. Chemical Analysis of Iron and Steel. P. 422.
22 L. Fine. *Metal Progress*, Feb., 1946.

Industrial Spectroscopy

THE Industrial Spectroscopy Group of the Institute of Physics, under the Chairmanship of Mr. F. Twyman, held its 1946 Conference in the Wellcome Research Institution, 183, Euston Road, London, N.W.1, on 5th and 6th July. The papers on the first day dealt with infra-red absorption spectroscopy. Dr. H. W. Thompson, of St. John's College, Oxford, spoke on "The Use of Infra-Red Spectroscopy in Analysis." Many complex organic mixtures are capable of being analysed, both qualitatively and quantitatively, by newer methods which make use of the pronounced difference in infra-red spectre arising from small differences in structure.

The types of apparatus available for this work, both in this country and in America, were described in the paper by Mr. G. F. Lothian, of the S.W. Essex Technical College, on "Equipment for Infra-Red Absorption Spectroscopy." Great advances have been made in recent years in such apparatus. Previously, infra-red methods were considered to lie in the province of academic research only, but the apparatus described has enabled the application of the methods in routine works practice. This has been achieved by the simplification of old apparatus and the production of new

apparatus designed to provide ease of manipulation coupled with increased accuracy.

Emission Methods

On the second day, the Conference devoted itself to emission spectroscopy. Messrs. A. S. Nickelson and F. W. J. Garton, of the Chemical Inspection Department, Woolwich Arsenal, described the application of the spectrograph to qualitative and quantitative estimation of metals, alloys, ores and inorganic substances generally. The former type of analysis is now a simple process, which is widely applied. Quantitative analysis, where the analyst is only concerned with minor constituents, is also thoroughly developed as a routine method. Where major constituents are the subject of the determination, however, the problem is not yet completely solved, and the accuracy is not yet as great in general as that to be obtained by chemical methods of analysis.

The choice of source is one which depends on the type of analysis demanded. Sensitivity is highest with the D.C. arc, but this suffers from low reproducibility. The reverse is the case with the simple A.C. condensed spark. Other sources developed from these, and recommended for various purposes, are the high voltage A.C. arc and the low voltage interrupted A.C. arc. A combination is provided by the Multi-source Unit, which can provide either an arc-like or a spark-like discharge, thus providing a choice according to the particular demands of the analysis.

Spectrographs

The medium quartz spectrograph is usually sufficient for non-ferrous metals, but ferrous and other metals with complex spectra usually require the large quartz spectrograph or the grating spectrograph. In photographing the spectrum, films give better results than plates, and there is also a lot to be said, in special cases, for direct photometry of the spectral lines. Photometry of plates, to complete a photographically recorded analysis, provides its own problems, and must be carefully carried out if large errors are not to arise.

Mr. A. Walsh, of the British Non-Ferrous Metals Research Association, in his paper, "Light Sources for the Spectrographic Analysis of Metals and Alloys," dealt in more detail with this aspect of the topic, with special reference to problems as yet unsolved. Mr. H. T. Shirley, of the Brown-Firth Research Laboratory, concluded the Conference with his paper, "A Statistical Examination of Sources of Error in the Spectrographic Analysis of Low-Alloy Steel." This paper also laid stress on the desirability of further investigation of reproducibility problems.

Miscellaneous Microchemical Devices

VII—A Device for Following the Washing of Precipitates

By J. T. Stock and M. A. Fill

THE removal of impurities from gravimetric precipitates by washing is of vital importance in analytical chemistry,[1] and is usually followed by withdrawing portions of the filtrate for test with a suitable reagent. The apparatus shown in Fig. 1 was developed for the study of the washing of precipitates in semi-micro analysis. It permits the testing of the filtrate at any desired stage of the operation, without the need of disconnecting or of interfering in any otherway. While designed especially for use under reduced pressure, it a'so operates when the filtration is carried out under gravity. A Pregl filter tube is shown in use in Fig. 1, but a micro-Gooch or Neubauer crucible can obviously be used instead. The inspection chamber A of the trap tube B retains a small quantity of filtrate, into which the reagent solution may be introduced from pipette C, thus allowing any reaction to be observed. Successive drops of filtrate rapidly flush out the inspection chamber, so that the test may be repeated when desired, until the absence of reaction shows the washing to be complete.

Fig. 2 shows the construction of the reagent pipette. The reservoir D contains the reagent, which, by operating rubber teat E, is drawn in or ejected by way of pinhole F. A small sphere G is formed by fusing the

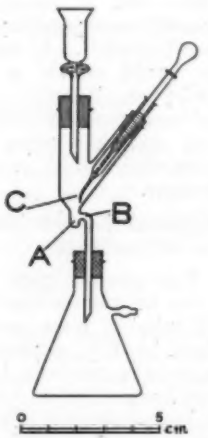

Fig. 1.—Apparatus for Washing Precipitates.

Fig. 2.—Reagent Pipette in detail.

extremity of the reservoir, and is ground lightly into the throat of the delivery capillary H. The reservoir slips loosely into the holder J, which is inserted into the side arm of the trap tube and connected with a rubber stopper cut from tube, as shown in Fig. 1. When it is desired to test the filtrate, a drop of reagent solution is ejected into J, and is delivered into the inspection chamber by a rapid up-and-down motion of the reservoir.

The device works very well with reactions involving a colour change, such as with acid-base or redox indicators, but is less satisfactory with precipitation reactions, e.g., when testing for freedom from sulphate with barium chloride solution.

ERRATUM: We regret that through an oversight the note, "A Device for Delivering Small Quantities of Liquids," which appeared in our last issue, was not attributed to its authors. This formed No. VI in the series, "Miscellaneous Microchemical Devices," by Messrs. J. T. Stock and M. A. Fill.

1 See, for example, T. B. Smith, "Analytical Processes," Arnold, London, 1940, pp. 61–64.

Chromium Steels of Low Carbon Content

By R. Franks

THIS investigation was carried out to determine the effects of chromium on such characteristics of low-carbon steel as resistance to corrosion and oxidation at elevated temperatures, and susceptibility to air hardening. The steels for the investigation were melted in high-frequency induction furnaces and were cast into ingots weighing up to approximately 50 lb. In melting, only manganese and silicon were used as deoxidisers to obtain sound steel. The finished steels contained from 0·15–0·35% silicon, 0·30–0·60% manganese, 0·025 and 0·035% phosphorus and sulphur maximum and, except in a few instances, carbon was limited to a maximum of 0·10%. All the constituents had to be controlled within those limits in order to determine the true effects of chromium. The small 2–3 in. square ingots of the different steels were first forged to small billets and subsequently rolled to ¾ or 1 in. rounds, and sheet or plate to obtain the samples for the various tests.

Twenty-gauge sheet samples of the steels containing from 0–17·26% of chromium in their annealed and descaled conditions were used in the atmospheric corrosion tests. These were exposed to the industrial atmosphere in Long Island City, N.Y., for several years and subsequently transferred to Niagara Falls, N.Y., where they were exposed for a period of about eight years. During the period of exposure the samples were cleaned after an exposure period of about seven years and again after an exposure period which varied between ten and eleven years. The oxidation tests at elevated temperatures were made on 1½ in. long sample cut from 1 in. round steel bars containing 0–23·61% of chromium. Machined samples, free from scale, were exposed to an oxidising atmosphere (air) at temperatures varying from 540°–800° C. for 900 and 1,000 hour periods. Susceptibility to air-hardening experiments were carried out on low-carbon steels, 0·06–0·098% carbon, containing 0–6·08% of chromium. These steels were tested in the as-hot-work condition, after softening by use of a subcritical annealing treatment at 725° C. and after air cooling from 900° C., Fig. 1.

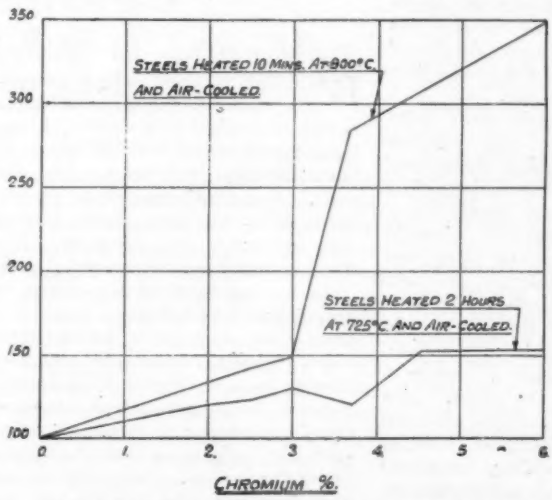

Fig. 1.—Effect of chromium on the air hardening of low-carbon steel.

The investigation has shown that wrought 3%-chromium steel containing up to 0·10% carbon, is at least five times as resistant to atmospheric corrosion in industrial atmospheres as ordinary low-carbon steel. At elevated temperatures the low-carbon 3%-chromium steel is superior to ordinary carbon steel. It possesses good resistance to oxidation at temperatures up to about 650° C., whereas ordinary carbon steel is considered unsuitable for temperatures in excess of about 540° C. Low-carbon 3%-chromium steel is also relatively free of air hardening, as shown by test on specimens air-cooled from 900° C., after which treatment the steel is soft, ductile and tough. Annealing at subcritical temperatures, 725° C., also softens the steel and imparts high ductility and toughness with a comparatively high yield strength at 0·2% offset.

Low-carbon 3%-chromium steels have good toughness at temperatures down to — 125° C. when previously given the subcritical annealing treatment at 725° C., but their toughness is not so high at the low temperatures when the steel has been previously normalised at 900°C. These steels have also good fabricating properties. According to the experience gained on small ingots, they can be hot-rolled in substantially the same manner as ordinary carbon steels. Their high ductility makes it possible to bend or otherwise form sections of the steel, either hot or cold, into any part produced from ordinary carbon steel. The 3%-chromium steel can also be welded and welds have a good toughness at room temperature and at temperatures down to — 40° C.

The utility of low-carbon 3%-chromium steel becomes apparent when its good fabricating properties and resistance to atmospheric corrosion are jointly considered. Articles of the steel, such as roofing, gutters, drainpipes, and other parts exposed to weathering, will last for long periods. Fence wire represents another article to be made from the steel, as well as certain structural parts for applications in which atmospheric corrosion is a serious menace.

From *Trans. Amer. Soc. for Metals*, 1945, Vol. 35, pp. 616–636).

Capillary of Metallic Surfaces

By E. R. Parker and R. Smoluchowski

THE spreading of liquids over the surface of solids is important in many engineering fields, and is of particular interest in many problems arising in brazing and coating. The behaviour of a liquid on a solid metallic surface depends entirely upon the condition of the surface. A critical survey of various theories of capillarity and of spreading of liquids on solids indicates that there is no simple and complete theory of the phenomena involved. In an attempt, therefore, to correlate and explain qualitatively certain observations made on the spreading of liquid metals on solid metals, consideration has been limited to surface energy and to some simple geometrical factors. No attempt has been made to go into the details of the atomic picture of the mechanism of the spreading of liquids.

Spreading of a liquid on a solid surface depends upon the relative magnitudes of the surface energies of the air-solid surface, air-liquid surface and solid-liquid surface. Spreading of the liquid decreases the air-solid surface and increases the other two surfaces. If the balance of these energy changes is negative, that is, if the total energy of the system decreases, then the liquid will spread otherwise it will tend to cover a minimum area of the solid. The balance of changes of the air-solid, air-liquid and solid-liquid energies depends upon the capillary roughness of the metallic surface and the roughness is approximated by a groove of an angle α inclined to the surface of the metal at an angle β.

Experiments were made with liquid silver on flat metallic surfaces subjected to various treatments and also on grooved surfaces in order to study the fundamental mechanism of spreading. The base metal used was iron containing 6% of molybdenum to prevent the occurrence of the α—γ transformation. In determining the effect of the angle α on the rise of the liquid in grooves, the iron sample had small V-shaped grooves 0·005 in. deep machined in the surface, and three angles 60°, 90° and 120° were used. The sample was placed in a hydrogen furnace with the grooves vertical and with a little silver at its base. It was then heated at 1,200° C. for 20 mins., when the silver melted and rose rapidly in the 60° grooves, slowly in the 90° grooves, and hardly rose at all in the 120° grooves. This behaviour was in agreement with the qualitative conclusions from theoretical considerations.

A special study was made of the influence of various surface treatments of the iron samples and for this purpose an area of about 5 sq. in. was treated by grinding, polishing and etching. By means of a profilometer the average roughness on the surface was determined, then a small disc of silver sheet was placed in the centre of each sample, the silver was melted in contact with the iron by heating in a hydrogen furnace at 1,200° C. for about 20 mins. and the sample cooled in hydrogen in a water-cooled chamber of the furnace. The results obtained from these tests showed that liquid metal spreads best on polished and etched surfaces, both of which have fine capillary structure, but that it does not spread on polished surfaces under the same conditions. On etched surfaces, spreading depends upon the etching reagent as well as upon the orientation of the grains.

From *Trans. Amer. Soc. for Metals*, 1945, Vol. 35, pp. 362–371).

Hard Chromium Plating Directly on Carbon and Graphite

By R. S. Herwig

BECAUSE of their unique properties, carbon and graphite products have been applied to a variety of uses including electrical brushes, bearings, liners, bushings, resistors, washers and seals. One disadvantage to carbon products is their comparative brittleness and low surface hardness. This disadvantage has been somewhat overcome by impregnating the carbon or graphite with metals or by combining the carbon or graphite powder with metal powders, compressing and then heating to the desired temperature. Recently, technique has been developed for direct plating of chromium on to carbon in order to increase the fatigue life of the carbon. Chromium plating also increases the rigidity of the carbon, has a dampening effect on the carbon, and imparts increased physical properties to the carbon or graphite without destroying any electrical or lubricating characteristics.

Initial attempts to plate chromium on graphite from a standard hard chromium plating bath resulted in no plating, or at best, a plate on the outer edges only. Increasing the voltage merely increased the amount of "gassing" without further deposition of chromium. Experiments were then directed to obtaining a plate from an intermediate bath such as copper and this proved more successful as good coverage was obtained in the rochelle copper bath provided the initial voltage was approximately 6–8 volts and good contact was made on the carbon. Subsequent deposits on the copper-plated carbon of nickel, cadmium or tin showed fair adhesion, but attempts to plate chromium on the copper base caused lifting. With a careful cleaning procedure, however, prior to copper plating, satisfactory adhesion was obtained. This cleaning procedure consisted of a soak in a mild alkaline bath followed by a short reverse current clean for a few seconds. The temperature of the bath was maintained at 70° C., as higher temperatures or a longer strike at reversed currents caused surface oxidation resulting in a poor bond.

Further experiments carried out in an effort to plate chromium directly on to carbon showed that it was necessary to accurately calculate the effective area. It was also necessary to use the greatest precaution in wiring carbon products for chromium plating due to the high resistance of carbon and the small contact area which resulted in an appreciable voltage drop between the wire and the carbon product being plated. Another important factor was that the wire holding the carbon product was completely "stopped off."

In plating, parts which have been securely wired and "stopped off" were immersed in a standard chromium bath at 45° C., containing 33 oz. of chromic acid per gallon and the ratio of chromic acid to sulphate maintained at 100 : 1. In order to obtain the highest current possible on the carbon product being plated, it was found necessary to move the anodes to within $\frac{1}{16}$–$\frac{1}{4}$ in. of the carbon part,

From *Materials and Methods*, 1945, Vol. 22, No. 6, pp. 1,751–53.

and this introduced another difficulty in that the placing of the anode so close to the parts being plated caused shorts and burning of the plate. For the best results, the carbon products should be placed on self-contained anode racks.

In applying current, the highest possible voltage available should be used as a strike. In the experiments carried out 8 volts was maintained for approximately 2 minutes or until complete coverage was obtained on the carbon product. This voltage was then decreased to 4 to 5 volts for 30 minutes and further decreased to 3 volts for 1½ hours. This combination of voltages resulted in a bright chromium deposit of approximately 0·002 in. thick, and this deposit could be cut and ground without any chipping of the bord between the chromium and the carbon. It is most important that the higher voltage should not be used for too long a period of time, as the chromium deposit that occurs at the higher current densities is more highly stressed due to the occluded hydrogen gas and chromic oxides which cause lifting and severe cracking of the chromium plate.

Experiments carried out on plating from a cold chromium bath indicated that the conditions would be less critical. The chromium deposit, however, was not as pleasing and contained greater porosity. Furthermore, the plating of chromium on carbon was performed with the thought of using the standard production chromium plating bath which operates at an elevated temperature.

Mechanism of Rusting of Low-Alloy Steels in the Atmosphere

By H. R. Copson

IN 1941, five sets of 71 low-alloy structural steels including high-purity iron with copper, copper steel with low phosphorus content, copper steel with high phosphorus content, copper steel with high manganese and silicon, nickel steel, nickel-chromium steel, nickel-molybdenum steel, nickel-chromium-molybdenum steel, nickel steel with copper, and nickel steel with copper and chromium, were exposed to the atmosphere at two locations, one being an industrial atmosphere, Bayonne, N.J., and the other a marine atmosphere, Block Island, R.I. Test specimens, 4 in. × 6 in., were mounted on metal racks supported on Monel frames by means of porcelain insulators. Data obtained in the course of conducting these tests has led to the development of a new mechanism of rusting.

An analysis of the rust from the two locations was made at the end of one year. There was more sulphate in the rust from the industrial atmosphere and more chloride in the rust from the marine atmosphere. Composite rust from 0·05% copper steels contained more moisture, more sulphate and more chloride than the rust from a fast corroding steel. Analyses were also made on rusts from selected steels after two years' exposure at Bayonne, when it was found that groundward surfaces of copper and nickel steels contained more moisture and sulphate than the rust from the skyward surfaces. Analyses of adherent rusts were of a similar nature but contained somewhat less sulphate than loose outer rusts. After three years' exposure, analyses were made on the loose rust collected from the skyward and groundward surfaces of a considerable number of copper and nickel steels exposed at Bayonne. The weight of loose and adherent rust was also determined and the approximate thickness of rust coatings obtained.

In addition to obtaining data from selected rust samples, an examination was made of weather effects with reference to edge corrosion and weight losses when it was found that correlation between weather data and weight losses was not so good as between weather data and edge corrosion. Certain relationships, however, were evident and had to be taken into account in formulating a mechanism of rusting.

From all the data obtained, it was concluded that corrosion rate depended on the quality and quantity of water reaching the steel surface. The quality was affected by atmospheric pollution, by the solubility of corrosion products, and by the wash of rain, while quantity was affected by shelter and by porosity of the rust which in turn was affected by the solubility of the corrosion products. Relatively insoluble corrosion products decreased the amount of the soluble constituents in the water and thereby decreased corrosion. Relatively insoluble corrosion products also did not wash away and thereby they decreased the porosity of the rust coating and the quantity of the water reaching the steel. As time went on pores became plugged and the position of low-alloy steel improved.

In industrial atmospheres, the function of copper and nickel in the steel was to render sulphate corrosion products more insoluble by forming complex basic sulphates. On mild steel, sulphates in rust were relatively soluble and promoted corrosion, but were washed away by rain. On alloy steels sulphates were less soluble so that corrosion was slower, but less sulphate was washed away and accumulated in the rust. The percentage of sulphates in the rust was found to increase as weight loss decreased. Porosity of rust also decreased with time, and rust on corrosion resistant steels was least porous. At the marine location, severe corrosion on certain edges was due to soluble chlorides being swept to these edges by the wash of rain.

Variations in weather affected corrosion, but different steels were differently affected. This was considered due to the rainfall having a dual role, in that moisture must be present for steel to corrode, but washing away of soluble matter was beneficial. Dew, fog, high humidity and shelter would also be expected to be harmful.

From *Amer. Soc. for Testing Materials*, 1945, Preprint 25, pp. 1–27.

Passivation of Stainless Steel

By E. M. Mahla and N. A. Nielsen

THE strength and stability of passive films produced on stainless steel by various passivating treatments have been evaluated. In addition, electron diffraction has been applied to the study of passivated stainless steel surfaces and of bulk oxide films. Break-through-potential data and confirmatory immersion-corrosion data showed that no passivation treatment confers lasting protection to stainless steel immersed in media which corrode the unpassivated metal. Equally effective results were obtained with stainless steel which was cleanly pickled, thoroughly rinsed, and air dried. Although no definite evidence of the existence of passive films was

From *Electrochem. Soc.*, Preprint, 89–27, April, 1946.

found by reflection electron diffraction studies, bulk oxides produced by chemical or air oxidation on 18-8-S stainless steel were found to consist chiefly of Cr_2O_3 with no nickel oxide and little iron oxide present. The bulk oxides formed by air oxidation on straight chromium-iron alloys consisted of FeO, Fe_3O_4, Fe_2O_3, and mixtures of these oxides.

Magnesium Cathodes for Pipeline Protection

By L. M. Oldt

FOR some time magnesium has been recognised as a desirable material for use in cathodic protection installations inasmuch as it combines in one metal a large storage capacity for electricity and a high potential or driving voltage, both desirable characteristics when the anode is to operate galvanically. Theoretically, the storage of electricity in magnesium is 1,000 ampere-hours per lb. This is a larger storage capacity than is realised with any other metal having a comparable driving voltage, with the exception of lithium. A further property of magnesium which recommends its use as an anode is the fact that it shows little or no tendency to polarise. It has been found that the current efficiency and life of magnesium anodes depend largely upon three factors, the composition of the metal used, the electrolyte used around the anode and the current density at which the anode operates. On an operating anode, two types of corrosion are in evidence. The first, or useful corrosion, results from the current provided by the anode for the protection of a remote cathode, such as a pipeline. The second, local, or non-useful corrosion represents current supplied by the magnesium to protect local cathodes in and on the magnesium anode itself. The ratio of useful corrosion to local corrosion determines the operating efficiency of the anode. A special alloy has been developed for anodes which realise a high current efficiency and reduces to a minimum the rate of local corrosion by controlling the copper, nickel and iron content. This alloy has the following composition: Al 5·3–6·7, Mn (min.) 0·18, Zn 2·5–3·5, Si (max.) 0·3, Cu (max.) 0·05, Ni (max.) 0·0003, Fe (max.) 0·003, other impurities (max.) 0·3%, Mg balance. Results of investigation on types of backfill show that gypsum-clay mixtures are more desirable than chlorides. It has been found that the most efficient backfill, to obtain high anode efficiency and life, is as shown in the following table:—

1. *For low-resistance soils:*
   3 parts Bentonite
   1 part ground Gypsum, Hydrated ($CaSO_4 \cdot H_2O$)
2. *For high-resistance soils:*
   2 parts Bentonite
   1 part ground Gypsum, Hydrated ($CaSO_4 \cdot H_2O$)
   1 part Anhydrous Sodium Sulphate.

As regards the effect of current density on anode efficiency, it may be said that when using the alloy, a current density of the order of 0·5 milliamperes/sq. in. is sufficient to obtain an efficiency of approximately 50%. In general, the anodes are placed about 10 ft. out from the line and are buried in augured holes 6–8 in. in diameter. They are connected to the pipe with suitable copper wire. Experience has shown that best results are obtained when the anodes are buried at such a depth that permanently moist soil is encountered. Spacing may range from 50 ft. to several miles. A 17-lb. anode supplying 1 ampere of current will have a life of about one year, while a 50-lb. anode at the same current flow will have a life of approximately three years. Reduced current output will prolong anode life. Special anodes have been designed by Dow Chemical Co. and are now available in three sizes. The anodes are all cast with a galvanised steel core and are furnished with 10 ft. of 14-gauge single strand direct burial copper wire.

National Petroleum News (U.S.A.), May 6, 1943. (From the *Abstract Bulletin* of Aluminium Laboratories, Ltd., 1946, No. 3).

The Effect of Working on the Physical Properties of Molybdenum

By J. W. Marden and D. M. Wroughton

MOLYBDENUM is commercially produced by powder metallurgy methods. The metal powder obtained by reducing the oxide MoO_3 or other compound in hydrogen is pressed to desired shape and heated to near the melting point in a hydrogen atmosphere to produce solid, coherent metal billets. These billets are forged and rolled or drawn at temperatures higher than those used for working steel. The physical properties of the finished molybdenum depend upon these temperatures and working schedules.

Commercial molybdenum powder has a particle size of 1 or 2 microns and, when lightly tapped in a cylinder or glass graduate, has a bulk density of about 3. The density of a pressed billet is normally 6 and of heat-treated ingots about 9·8. The density of the heat-treated pieces, however, depends somewhat upon the pressures used for compacting. Very high pressures on thin sections result in densities greater than 10. When the heat-treated ingot is worked into sheet metal or wire, however, there is a gradual increase in the density.

The hardness of commercial molybdenum also increases with the amount of working, but, although there is a general trend toward greater hardness as the cross-sectional area decreases, there is at any one thickness a considerable variation in the Vickers pyramid number. The rate at which hardness develops depends upon the temperature as well as upon the size of the reduction step during the working process.

As with hardness, the ultimate tensile strength of molybdenum shows a gradual increase from 50,000 lb./sq. in. for unworked metal to 200,000 lb./sq. in. for fine wire or very thin sheet. There is also, at any one cross-sectional area, considerable variation from the mean.

Pure heat-treated unworked molybdenum can be deformed to some extent at room temperature by slow application of pressure, but it shatters with sharp impacts. After suitable working, sheet metal or wire becomes ductile and may be cold deformed into useful shapes. After a reasonable amount of working has been done on the metal, the per cent. elongation on 2 in. may vary from a few per cent. to 25. Special samples have given values up to 50%. The values do, however, generally increase with the reduction in cross-sectional area.

From *Electrochem. Soc.*, Preprint, 89–30, April, 1946.

Hardness, strength and ductility at room temperatures are not inherent in molybdenum but are developed by proper working and heat-treatment. Samples of well-worked hardened molybdenum heated for 5 mins. at various temperatures begin to lose their hardness at 1,000° C. In longer periods of time or at higher temperatures, the hardness returns to the original value of crystallised unworked metal. Recrystallisation occurs and the microstructure reverts to the equiaxed structure. Annealing, however, may improve the properties of some samples. It appears that severely cold-worked samples may have better strength and yield after proper annealing, but when over annealed they are not unlike the original treated ingots.

Tensile Properties of Aluminium Sheet at Elevated Temperatures

By A. E. Flanigan, L. F. Tedsen and J. E. Dorn

IT is necessary occasionally to use aluminium-alloy sheet where moderately elevated temperatures are encountered. Investigations on the properties of aluminium sheet at elevated temperatures are few, however, and as they have not always covered the range of times and temperatures of greatest interest, the present investigation was instituted to overcome this deficiency. For the purpose of the investigation it was found convenient to classify the high-strength alloys into two groups. Group A included precipitation-hardenable alloys that had been solution-heat-treated and aged at room temperature (naturally aged), and Group B, precipitation-hardenable alloys which had been solution-heat-treated and aged at elevated temperatures (artificially aged). This classification permitted obvious generalisations on the response of each group to further elevated-temperature exposure.

The alloys of Group A consisted of 0·040 and 0·064 in. 24 S-T and 24 S-RT Alclad sheet and 0·040 in. 61 S-W bare sheet, and those of Group B of 0·040 in. 24 S-T80, 24 S-T81, 24 S-T86, XB 75 S-T Alclad sheet and 0·040 in. 61 S-T bare sheet. Alloy 24 S contained approximately 4·5% copper, 0·6% manganese and 1·5% magnesium; alloy 61 S, 0·25% copper, 0·6% silicon, 1·0% magnesium and 0·25% chromium; while no precise information was available for the composition of 75 S, although the principal hardening elements were zinc and magnesium with smaller additions of copper and other elements.

Tensile tests were carried out on each alloy to determine: (1) Ultimate tensile strength; (2) tensile yield stress; (3) elongation; (4) primary and secondary moduli of elasticity; (5) tangential moduli for all stresses up to the tensile yield stress. Results at room temperature are given in Table I. All tests were made with a crosshead speed of 0·06 in. per min. and with specimens stressed transversely to the direction of rolling. Temperatures investigated included 25°, 100°, 150° and 190° C. At each of the elevated temperatures exposure periods ranged from ¼ hour to 1,000 hours.

Tensile tests at elevated temperatures showed the changes in properties to be produced either by the effect of temperature alone independent of attendant precipitation effects, or to result from structural changes in the material induced by elevated temperature exposure. In the case of the shortest exposure time (¼ hour), the principal effect was that of temperature alone, increases in temperature producing lower values of yield stress and ultimate stress. With continued exposure, especially at higher temperatures, structural changes occurred and the materials of groups A and B behaved in dissimilar manners. The effects noted in the specimens of group A were those of marked precipitation-hardening followed in some cases by overageing. The tensile properties of the Alclad materials were little affected by exposures of 1,000 hours at 100° and 120° C. At 150° C. there was a substantial increase in yield stress with eventual overageing, whereas early overageing was apparent at 190° C. At the latter temperature, the yield stress after 1,000 hours was scarcely lower than the ¼-hour value but there was a marked drop in ultimate tensile stress. In the 61 S-W material similar trends were apparent but precipitation effects were more prominent at the lower temperatures. Since the materials of group B had been subjected to previous artificial ageing the effects were generally those of overageing, the values after 1,000 hours being similar to the ¼-hour

TABLE I.—TENSILE PROPERTIES OF MATERIALS AT ROOM TEMPERATURE.

Material	Tensile Yield 1,000 lb./sq. in.	Ultimate Stress 1,000 lb./sq. in.	Elongation % on 2 in.	Primary Modulus 10^6 lb./sq. in.	Secondary Modulus 10^6 lb./sq. in.
24 S-T Alclad	43·3	64·8	17·0	10·4	9·2
24 S-RT Alclad	52·3	67·0	10·6	10·6	9·7
24 S-T80 Alclad	51·6	65·0	10·0	9·9	9·0
24 S-T81 Alclad	58·9	65·2	5·3	10·4	9·3
24 S-T86 Alclad	66·6	70·4	4·5	10·3	9·1
61 S-W Bare	23·2	36·8	10·2	10·0	—
61 S-T Bare	38·3	44·1	12·0	10·1	—
XB 75 S-T Alclad ..	67·0	78·0	10·3	9·9	8·8

TABLE II.—EFFECT OF TEMPERATURE IN THE MODULUS OF ELASTICITY IN TENSION.

Material	Modulus	Temperature 25° C.	100° C.	120° C.	150° C.	190° C.
24 S-T Alclad	Primary	10·4	9·6	9·6	9·4	9·2
	Secondary	9·2	8·7	8·7	8·5	7·9
24 S-RT Alclad	Primary	10·6	10·2	10·1	9·8	9·2
	Secondary	9·7	9·2	9·2	8·8	8·1
61 S-W Bare	Primary	9·9	9·1	9·1	8·9	8·3
	Secondary	—	—	—	—	—
24 S-T80 Alclad	Primary	9·9	9·4	9·4	9·2	8·7
	Secondary	9·0	8·8	8·6	8·4	8·0
24 S-T81 Alclad	Primary	10·4	9·7	9·6	9·7	9·1
	Secondary	9·3	8·9	8·8	8·6	8·1
24 S-T86 Alclad	Primary	10·3	9·7	9·7	9·5	8·6
	Secondary	9·1	8·9	8·7	8·4	7·6
61 S-T Bare	Primary	10·1	9·3	9·2	8·8	8·3
	Secondary	—	—	—	—	—
XB 75 S-T Alclad	Primary	9·9	9·3	9·2	8·9	8·2
	Secondary	8·8	8·5	8·4	7·8	—

From *Metals Technology*, 1945, Vol. 12, No. 8, and A.I.M.M.E. Technical Publication No. 1929, pp. 1–19.

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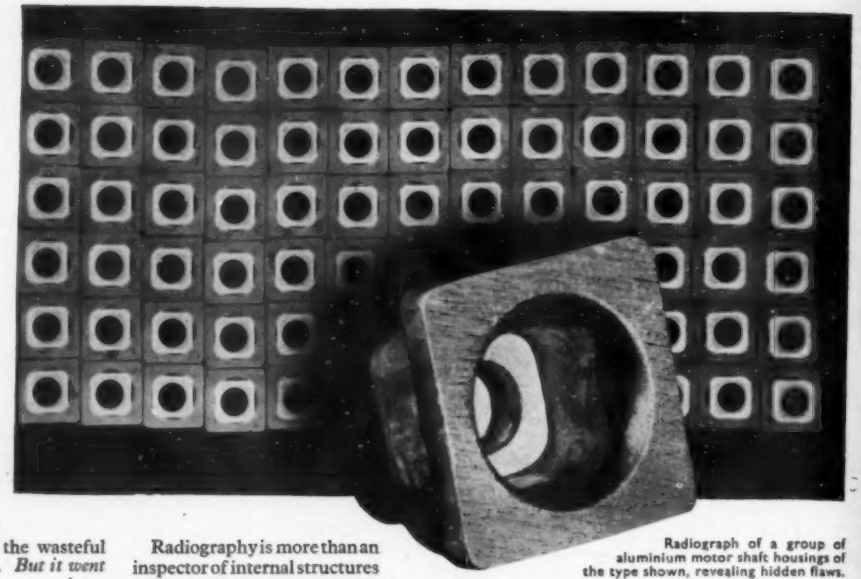

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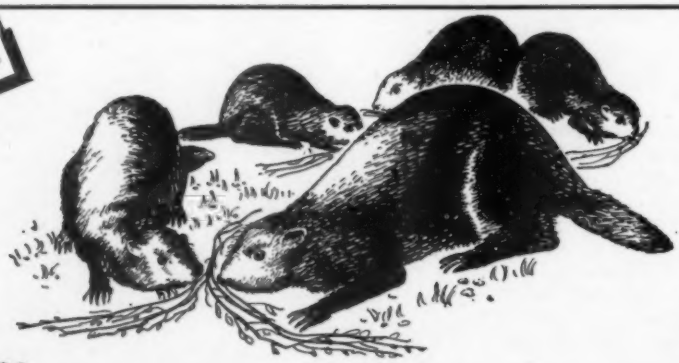

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values. There were considerable decreases in both yield stress and ultimate stress at 190° C. and smaller decreases at 150° C. The greatest loss in strength occurred in XB 75 S-T, which had the lowest temperature for artificial ageing.

The occurrence of a secondary modulus of elasticity in clad aluminium alloys is known to be due to the dissimilar characteristics of the cladding and core materials, the former having a relatively low yield stress, and secondary moduli were observed in all clad materials except where prolonged exposure at elevated temperatures had caused a large decrease in the yield stress of the core material. Values of the primary and secondary moduli decreased with increasing temperature. The data summarised in Table II were treated as dependent on temperature alone, since no effect of exposure time was noted. Tangent modulus curves at elevated temperature showed that where time was a factor families of curves were obtained. As with other materials, creep was a significant factor in elevated temperature applications where aluminium alloys sustain loads for long periods and stress-rupture curves at 150° and 190° C. have shown that eventual failure by rupture may occur under steady loading at stresses well below the value of the short-time "tensile yield stress."

In consideration of aluminium alloys for use at elevated temperatures the problem of corrosion resistance requires attention. In 24 S-T and 24 S-RT, a marked increase in susceptibility was found to occur in the early stages of elevated temperature ageing. With continued ageing corrosion resistance improved and approached that of the original material and the ageing time required for optimum resistance coincided with that required for attainment of maximum yield stress. As a consequence, corrosion problems may be expected to attend the use of those alloys at 150° C. for under such conditions considerable time is required for attainment of maximum yield stress. Such difficulties may be minimised through the use of the 24 S-T8 series. The corrosion resistance of artificially aged XB 75 S and 61 S have also been found adequate after the period required to attain maximum yield stress and so render them suitable for elevated temperature service.

Charging Basic Electric Furnaces

By Henry Bigge

THE relative merits of top charging versus side door charging in the larger types of basic electric furnaces remains a moot subject. However, most of the recent installations of 50-tons capacity or more have been of the door-charging type. The chief problem is one of fitting the unit to the plant layout.

The top-charging method consists of two designs: the gantry roof lift type and the roof swing type. In the former design the electrode masts and the roof raising equipment are built into a gantry crane which travels on rails along the charging floor. When the furnace is to be charged, the electrodes are raised to clear the shell and the roof is lifted. The gantry is then moved aside, carrying the roof and the electrodes; all movements are motor operated. When the gantry is not being used to remove the roof it is locked to the furnace and tilted with it. In the roof swing type the roof and the supporting structure for the electrode masts are lifted and swung to one side by a vertically operated oil-hydraulic ram and cylinder. Incorporated in this cylinder is a horizontal double-acting cylinder having a quadrant member so designed that the vertical ram may be turned in either direction only when the ram is in the maximum lifting position. A three-way piston type valve operates the vertical cylinder to lift the roof. Roof swing is controlled by a decelerating valve which controls the speed of the swing to maximum positions, open or closed. This valve decelerates to the point of complete shut-off, thus preventing shocks at the end of the swings.

The side door method consists of: chute charging, auto-floor charging, and machine charging. Chute charging has been widely employed for light miscellaneous scrap, such as is utilised in foundries. The chute is designed to take part of the charge of average-sized scrap. To facilitate the sliding of the scrap, the chute is fitted with pneumatic or electric agitators. The furnace is tilted about 15° and the chute is suspended from the overhead crane, and inserted through the charging door opening. At present it is in use on a 15 ft. shell-diam. furnace of 40 net tons capacity.

The auto-floor charging machine is self-contained and requires no track or runways. It is electrically operated; the power for the two motors is transmitted through a flexible cable attached to a rotating collector which is mounted at the top of a mast. The movement of the machine is not restricted to any definite path and it may thus operate freely at any point around the melting floor, limited only by the length of the cable. A good floor is needed for the successful operation of this machine. It can be used in congested areas because it turns on its own wheel base. This machine is equipped for handling small charging boxes similar to the method employed in the machine charging used in the open hearth shops.

The machine charging device is an electrically - operated manipulator mounted on rails on the melting platform similar to the machine used for charging scrap into open hearth furnaces. It is equipped with a charging reel for handling small boxes of scrap; the size and capacity of the boxes depend on the opening in the charging door.

From paper presented at Conference of Iron and Steel Div. of A.I.S.M. at Cleveland, Dec., 1945.

Electrical End Welding Process for Stranded Wires

By H. Gunther

TO join electrical conductors made from stranded wire, in addition to the mechanical process of cramping and splicing, soft and hard soldering processes are employed to a considerable extent, and in the case of aluminium cables, also, gas welding and cast welding processes. Now, an electrical and welding process has been developed, which is arranged to operate without flux or addition materials and requires no post-working of the welding joint, because no welding protrusion is formed. By this means, the clean bare ends of the wires are so tensed in the welding apparatus that they are arranged to butt together in the middle of a cylindrical, ceramic chamber, in an endwise manner; the bore of the chamber is arranged to be some 10% greater than the diameter of the stranded wire. A spring that is capable of regulation, on one of the movable tensing jaws, presses the ends

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of the wire strands together, with a constant pressure. By means of a short current impulse from a transformer, the two ends of the stranded wire are melted by actual resistance heating; with continued melting, one of the strand ends is pushed back in an axial direction, until the complete cross-section of the ceramic chamber is filled with molten material. The current is then switched off by means of a contact, controlled by the apparatus pushing back the wire.

An apparatus arranged to work according to the above principle, for the welding together of copper wire, 1·5 sq. mm. has proved fairly perfectly satisfactory in workshop operation. The actual duration of the welding period with this equipment amounts to 0·3 secs. The reduction in the mechanical strength, caused by the heat-treatment of the wire during the welding process, must generally, be taken into consideration. Further experiments have shown that this equipment is quite suitable also for welding stranded wires of Aldrey, bronze, steel-copper and steel; in the same manner, it is quite feasible also to weld solid circular conducting wires to stranded wires. Experience that has been gained up to now with the process, has shown that for non-metals, the most favourable current strength to employ, amounts to 800 amps./sq. mm., and for steel the figure is 1,300 amps./sq. mm.

Cast-to-Size Impact Specimens for Aluminium Sand-Casting Alloys

By R. A. Quadt

THE usual procedure for obtaining specimens for impact testing of metals cast in sand involves the machining of the finished specimen from an oversize cast-to-shape test specimen, or the machining of a specimen from a section of a production casting. Tests were carried out to investigate the possibility of producing cast-to-size impact test specimens in green sand to satisfy the following conditions: (1) Multiple test specimens from one casting required so that a number of specimens would be available for duplicate tests; (2) test specimens from a single casting to produce reasonably consistent impact values; (3) test specimens from different castings produced under similar conditions to give comparable results. To predict how the impact value obtained would compare with those obtained on specimens prepared by standard practice, machined test specimens were used for comparison purposes throughout the investigation.

The type of specimen investigated was the 10 by 10 by 55 mm. Charpy test specimen. Wherever a notch was provided, it was the V type conforming to A.S.T.M. specifications. A commercial aluminium 4% copper alloy, melted in a crucible furnace, was poured at 660° C. into a group of 26 unnotched test specimens. Two specimens were tested in the as-cast condition four days after they were cast, another set was tested in the solution-heat-treated condition and a fourth and fifth set tested after various ageing treatments. Solution heat-treatments were conducted at 515° C. for 15 hours, the specimens being aged 24 hours at room temperature before testing or artificially aged. Ageing at 155° C. was carried out in constant temperature, electrically-heated salt baths. The results obtained, summarised in Table I, showed that no significant effect due to position in the casting could be determined from the individual results.

To determine the effect of notches in cast-to-size specimens, the pattern was altered so that one half of the specimens had V notches cut into the drag surfaces. The notched bars from one of the castings were tested about 6 hours after pouring, while the notched bars from other castings were heat-treated and tested as before. The results obtained are also summarised in Table I.

To obtain comparable values on standard machined and notched specimens, ten test specimens were cut from the bottom of the outside runners where the feeding and freezing characteristics most closely approximated those of the cast-to-size specimens. Six more specimens were cut from similar locations in other castings and milled 10 mm. square but left unnotched. Finally, notches were machined into 15 of the cast-to-size bars. All were heat-treated and tested as previously. The results obtained are given in Table II.

The results of the investigation indicated that it was possible to produce quickly and cheaply cast-to-size Charpy impact specimens to satisfy the conditions investigated. Both notched and unnotched specimens can be obtained from a single sand casting, and it was thus possible to obtain data on two sometimes unrelated properties of an alloy, its impact resistance and its sensitivity to notching. For the aluminium alloy tested, the values obtained from cast-to-size notched specimens closely checked those obtained from bars machined and notched from a comparable oversized casting. Unnotched cast-to-size specimens, however, have approximately double the impact values obtained from unnotched machined specimens.

The method of producing impact specimens investigated has been incorporated into a research programme with very satisfactory results, and several thousand specimens have been tested and cast to date on most of the common aluminium casting alloys.

TABLE I.—IMPACT STRENGTH OF CAST-TO-SIZE SPECIMENS.

Heat Treatment	Unnotched Specimen		V-notch Cast Into Specimen	
	No. of Specimens	Ft.-lb. Mean	No. of Specimens	Ft.-lb. Mean
AS-CAST :—				
First Casting	25	8·4	13	2·1
Second Casting	21	8·2	—	—
Solution heat-treated	31	18·5	9	2·7
Solution heat-treated and aged 3 hours at 155° C.	14	13·4	7	4·6
Solution heat-treated and aged 6 hours at 155° C.	13	12·3	7	2·0
Solution heat-treated and aged 24 hours at 155° C.	13	7·3	7	1·4

TABLE II.—IMPACT STRENGTH MACHINED.

Heat Treatment	Machined and Unnotched	Machined and V-notched	Cast-to-Size with machined V-notch
As-cast	4·0	1·4, 1·3	1·8, 1·8, 2·0
Solution heat-treated	11·0, 9·7	4·1, 4·7	5·2, 5·7, 5·4
Solution heat-treated and aged 3 hours at 155° C.	7·6	3·0, 3·2	3·0, 3·4, 3·2
Soution heat-treated and aged 6 hours at 155° C.	4·7	2·2, 2·5	2·6, 2·4, 2·8
Solution heat-treated and aged 24 hours at 155° C.	2·7	1·3, 1·4	1·5, 1·5, 2·0

From *Bulletin Amer. Soc. Testing Materials*, 1945, No. 135, pp. 24–26).